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CREEP RATES FROM TESTS OF SHORT DURATION

BY J. J. KANTER AND E. A. STICHA

Abstract

A number of papers have been presented dealing with the subject of short duration tests to serve the purpose of verifying materials for elevated temperature use. It is a serious problem among testing engineers to develop some suitable acceptance tests for materials to be used at elevated temperatures which reliably indicate good creep properties. Obviously, the execution of actual creep tests is such a lengthy and expensive process that as such they cannot serve for acceptance testing purposes. Clark and White have proceeded upon the basis that a relationship exists at a constant temperature between the long and short-time rupture properties of steels, studied with load as the variable. In the present paper a correlation is suggested from another standpoint which may for some purposes be more convenient. It is attempted to show that for a given load condition a relationship between short and long-time creep rates can be established with temperature as the variable. For various theoretical reasons it is held that the temperature variable is a much more fundamental one to study and possibly affords more promise in establishing a reliable long-time short-time correlation than does the load variable into which much more complex influences enter.

IN designing equipment for elevated temperature service, creep is one of the many factors that must be considered. The permissible deformation will dictate, to a great extent, the means of obtaining reliable design data. Installations in which the metal is operated up to the point of actual failure, as in oil still tubes, may be designed adequately from stress-rupture data. On the other hand, in the design of turbines, depending as they do upon the maintenance of small clearances over long periods of time for economical operation, due regard must be given primary creep which may constitute a large portion of the plastic deformation. In many instances, as in pressure

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vessels, the stress which will produce a particular secondary creep rate at the temperature in question is the basis for design, and creep in the primary stage is considered of negligible significance. It is the aim of this paper to present a short-time test method which will yield creep data of this last type.

The tensile creep test in which extension measurements are made over a prolonged period of time, usually 1000 hours or longer, on specimens under constant load and at constant temperature is the usual means of obtaining data for this purpose. The common procedure is to test a series of specimens at constant temperature under a series of loads. Thus, the precise measurement of the small creep rates, which are encountered in practice, is an expensive and time consuming task requiring considerable care and special technique.

DISCUSSION OF POSSIBLE METHODS

Many attempts have been made to short-cut the long-time creep test. In general, these attempts may be divided into two classes. In the first class are those methods based on arbitrary limits set up by their authors, presumably from their experiences with laboratory tests and actual installations. In this class we find Hatfield's "Time-Yield" method (1),¹ the method of Pomp and Dahmen (2), Jasper's "Long-Time Proportionality" method (3), and others. A discussion of the various methods proposed up to 1931 may be found in a book by Tapsell (4). Application of short-time tensile data for design purposes within the creep range of metals would fall in this class inasmuch as its success would depend largely on the proper choice of a safety factor. While all of these methods have been used with more or less success, primarily by their authors, they leave much to be desired. Long-time creep test data or its equivalent appears to be a more rational design basis.

Comprising the second class are those methods employing extrapolation, either of creep curves, secondary rates, or fracture times. In extrapolating creep curves, the aim is to obtain the total creep rather than the secondary creep rate and these methods will not be considered here. They have been discussed by Weaver (5), Kanter (6), and Marin (7). Numerous equations and plots have been proposed for the interrelation of rate, stress, and temperature, most of them endeavoring to correlate rate and stress at constant tempera-

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

ture. Any device which will yield a straight line plot over a wide range of rates has possibilities as a short-time test method. In this group we find Norton's log-log representation (8), Ludwik's semi-log correlation (log of rate) (9), and Kanter's relationship between stress and log flowability (rate/stress) (6). While the desired straight line correlation may be obtained for a particular set of data by the use of one of these devices, over a wide range of rates, departure from a straight line is generally considerable. Hence, only a limited amount of extrapolation is permissible and obtaining small creep rates from short tests is a questionable procedure. This matter has been discussed more fully in the A.S.M. Metals Handbook (14).

More recently, Kanter (10) proposed a relationship between the logarithm of creep rate and the reciprocal of the absolute temperature at constant stress. In this analysis, creep is compared to self-diffusion processes as observed in metals. A straight line plot of the creep rates is obtained, the slope of which may be interpreted as an index to the threshold energy that is required for the diffusion of the particular crystal fragment or domain involved. All steels investigated give lines of about the same slope, their position differing only with respect to the temperature scale. This method suggests a means of correlating small creep rates with large creep rates which has certain advantages over those previously proposed, and the intention of this paper is to suggest and demonstrate its possibility. The principle of fixing the stress and making temperature a variable appears a more reasonable attack in view of the success of such thermal analyses in the field of physical chemistry. The physical concept of thermal effects upon creep is perhaps more concise than that of loading effects, which undoubtedly are extremely complex. It seems therefore that temperature would be the simpler variable to attempt to correlate.

It should be mentioned that the method employing extrapolation of fracture times and the stress-rupture correlations proposed by Clark, White and others (11) have been seriously considered as acceptance tests. At constant temperature, plotting stress against fracture time on logarithmic co-ordinates yields a straight line plot and unless the metal is susceptible to intergranular oxidation or corrosion, this line may be determined quite accurately from a few relatively short tests. Parker (12) has shown some of these results correlated in the same manner as proposed by Kanter for creep rates

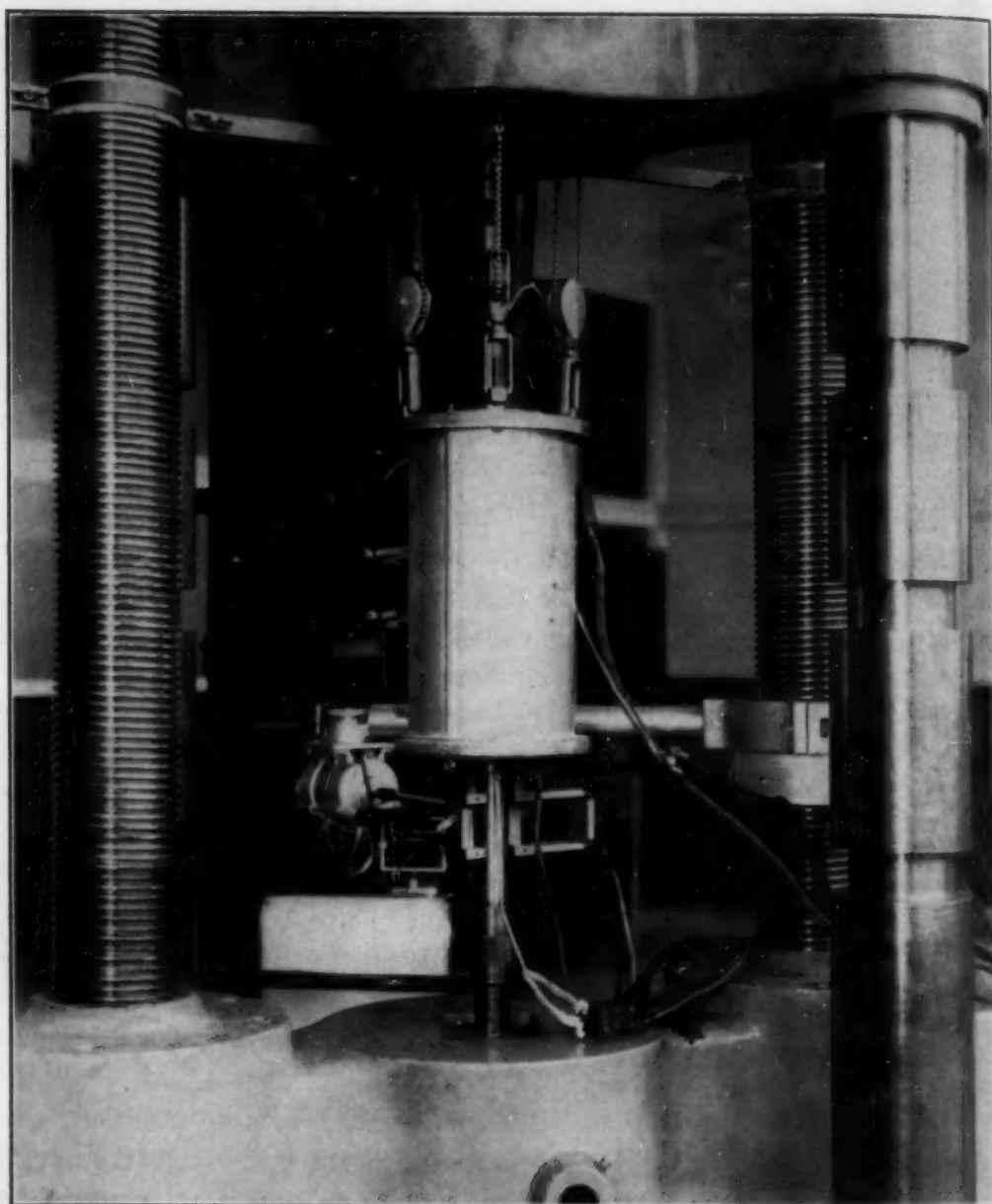


Fig. 1—Apparatus for Short-Time Elevated Temperature Tensile Test Showing Electrical Extensometer.

by plotting the logarithm of fracture time against the reciprocal of absolute temperature.

APPARATUS AND PROCEDURE

While tests were conducted primarily for the purpose of checking the possibility of extrapolation by Kanter's scheme, some were made to obtain data for checking other methods. All creep tests and tests of more than 50 hours' duration were made in creep units,

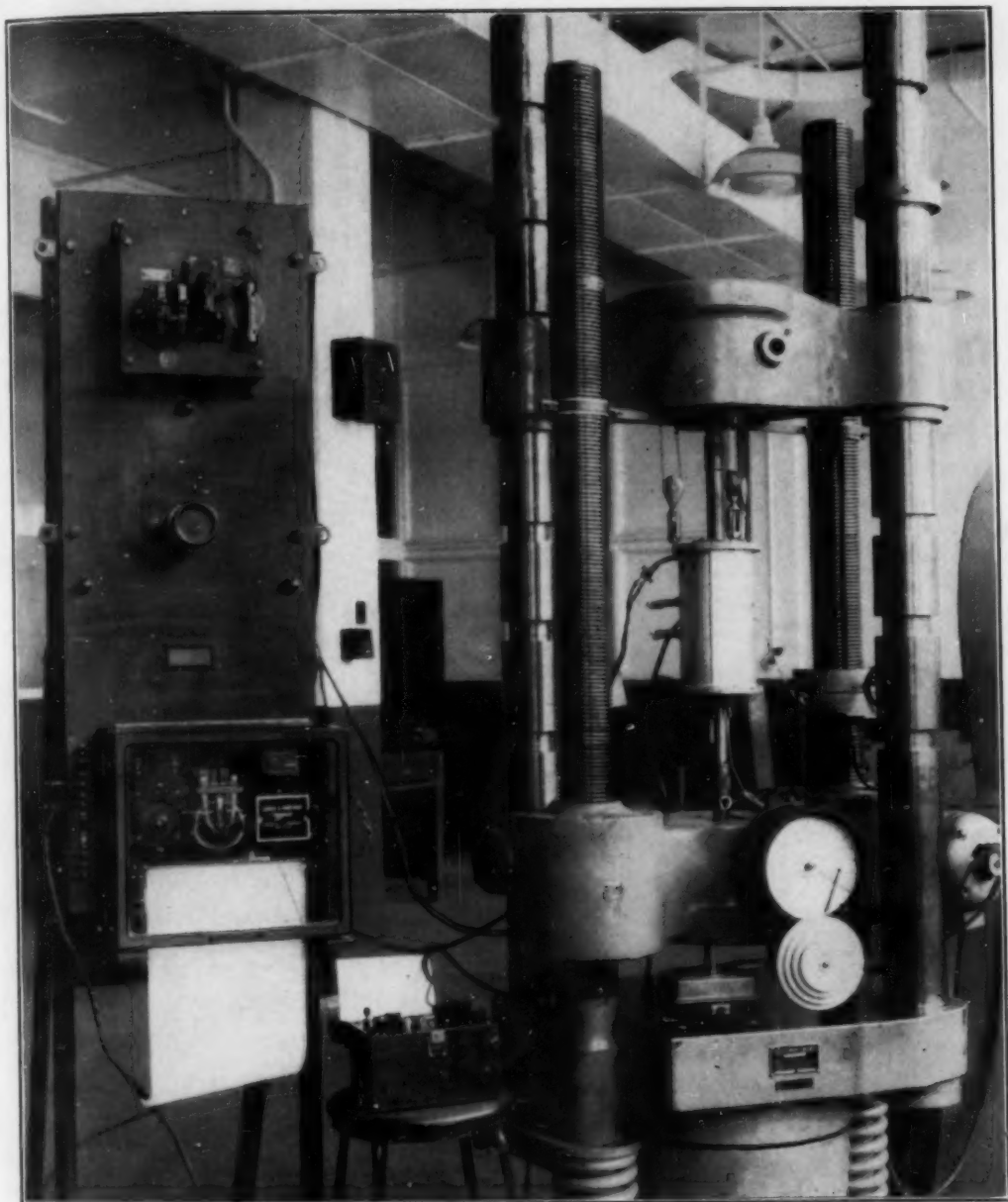


Fig. 2—Apparatus for Rupture Test Showing Crosshead Motion Recorder.

the construction of which has been described in detail elsewhere (13). Test bars are 0.505 inch in diameter with a 2.3-inch gage length, and they are loaded by weights acting through a 9:1 lever and threaded holders. An electric resistance furnace, wound noninductively and equipped with a metallic sleeve and taps for adjusting temperature distribution, supplies heat to the holders and test bar. Each unit is controlled by a potentiometer controller which maintains the temperature within ± 3 degrees Fahr. of the test temperature

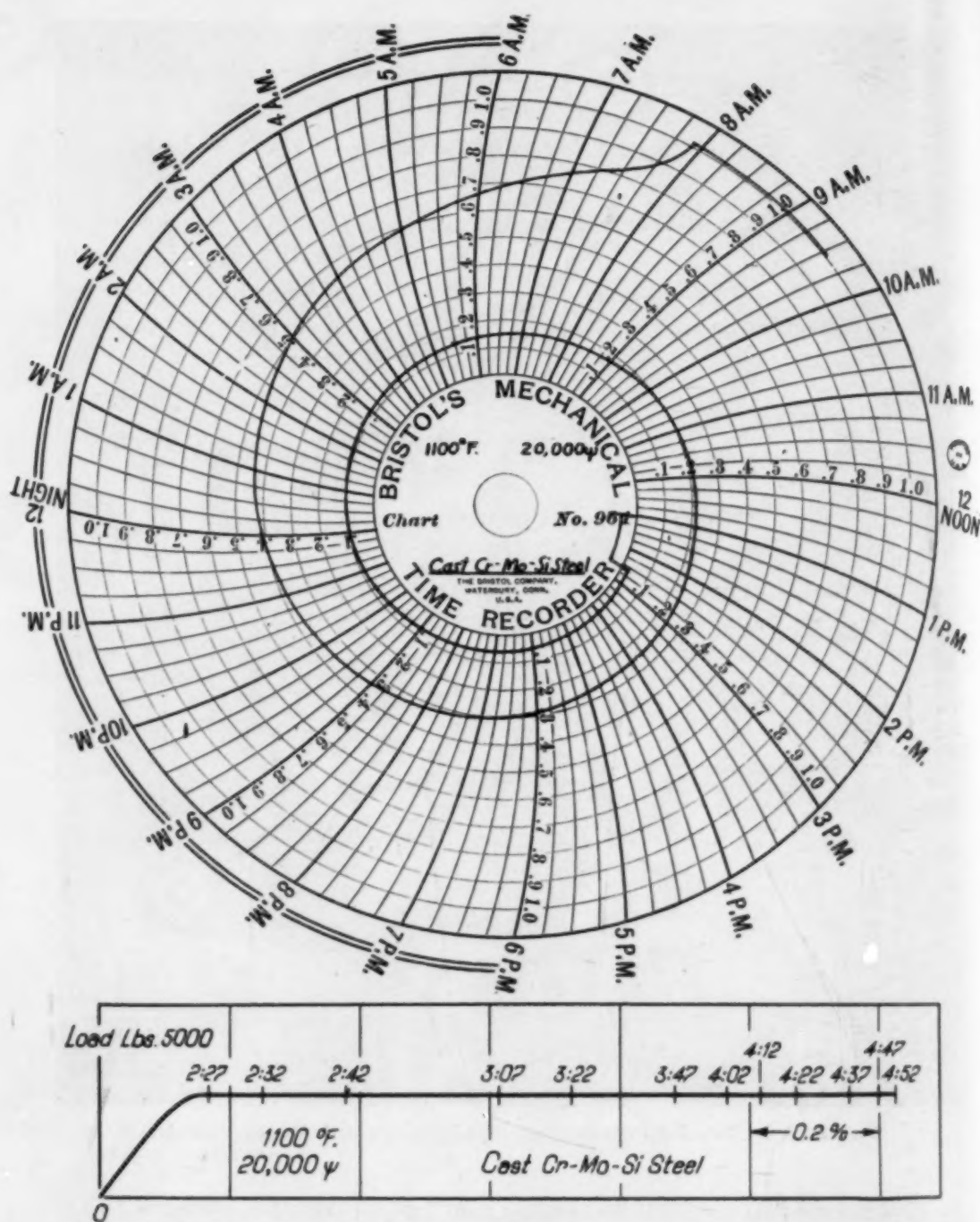


Fig. 3—Typical Curves from Cross-Head Motion Recorder and from Automatic Stress-Strain Recorder at Constant Load for Cast 5 Per Cent Cr-Mo-Si Steel.

while a multiple point recorder keeps a running record of the temperature. Extension measurements are made with a micrometer telescope equipped with a filar eyepiece capable of duplicating readings to 0.00005 inch by sighting onto platinum strips attached to the ends of the gage section.

Short tests were conducted in a hydraulic testing machine equipped with a load maintainer and a stress-strain recorder. The machine in effect becomes another creep unit and is shown in Figs. 1 and 2 as set up for these tests. The same test specimen is used and to it are attached arms which actuate an electrical extensometer suspended from the arms below the furnace. Straining the specimen causes a pen to be driven across a drum while the application of stress causes rotation of the drum. After the desired load is being maintained, time intervals may be indicated on the constant load line at the indicated strain. The extensometer system has a 2 per cent range, and its sensitivity is about the same as that of the creep telescope. For registering additional motion and fracture time, a time-motion meter is mounted on the fixed crosshead of the machine and actuated by the movable head. Inasmuch as the crosshead movement at constant load is due entirely to elongation of the specimen, this represents another means, although a less accurate one, for obtaining the time-strain curve in short tests. Fig. 3 shows the time-strain curves obtained by the two methods. The effective gage length, in this case, is obtained from a measurement of the elongation in the test section and the chart range covered. Good agreement is obtained by the two methods, there being usually less than a 10 per cent difference between rates. Temperature control is the same as on the creep units. The furnace is constructed in the same manner as the creep furnaces but is somewhat larger in diameter. It is suspended so that it moves at one half the crosshead speed, always maintaining the test bar in a position symmetrical to the furnace, thus minimizing the error due to temperature drift as the specimens elongate and the holders withdraw from the hot zones of the furnace.

All tests were made under conditions of constant temperature and constant load, some creep rates being obtained by step tests as indicated in Table I. Tests in which the bar is fractured in a relatively short time have been termed rupture tests by Clark and White to distinguish them from the usual creep test and this terminology will be used when necessary even though fracture time was merely incidental in this investigation. In rupture tests, the time-strain curve was the important feature, particularly the constant rate portion. These curves may be divided into three stages in the same manner as the usual creep curve and a further analogy is the fact that the final stage is entered usually after approximately 3 per cent creep. Thus, the electrical extensometer with its 2 per cent range

Table I
Cast Chromium-Molybdenum-Silicon Steel

Analysis:						
C	Cr	Mo	Si	Mn	S	P
0.23	4.96	0.50	1.08	0.66	0.008	0.028
Heat Treatment:						
Normalize 1800 degrees Fahr., air quench 1600 degrees Fahr., draw 1275 degrees Fahr.						
Short-Time Physical Properties:						
Temp. °F.	Tensile Strength P.s.i.	Yield Stress P.s.i.	Prop. Limit P.s.i.	Elong. in 2" %	Red. of Area %	Modulus of Elas. P.s.i.
75	97,300	68,600		25.0	59.0	
300	86,300	59,900	52,900	26.5	63.3	
550	83,800	59,300	50,600	23.0	58.9	27.5x10 ⁶
800	75,900	58,000	34,600	19.5	50.1	26.6x10 ⁶
1000	58,100	45,800	19,800	27.5	65.7	22.9x10 ⁶
Creep and Rupture Characteristics:						
Temp. °F.	Stress P.s.i.	Secondary Rate In./In./Hr.	Duration of Test Hours			
1000	40,000	0.026	3.12—Fractured			
800	30,000	0.00000039	1344.			
900	30,000	0.000047	624.			
1000	30,000	0.0049	22.1 —Fractured			
1100	30,000	0.0354	1.98—Fractured			
900	30,000	0.0000407	192.			
1000	20,000	0.000288	25.4			
1100	20,000	0.00316	41.3 —Fractured			
1200	20,000	0.102	0.7 —Fractured			
950	10,000	0.00000050	1992.			
1050	10,000	0.0000266	480.			
1200	10,000	0.00248	65. —Fractured			
1050	10,000	0.0000153	720.			
950	10,000	0.00000042	1008.			
1100	10,000	0.0000654	1272. —Fractured			
1300	10,000	0.0348	7.1 —Fractured			
1100	3,000	0.00000076	1512.			
1100	1,000	0.000000054	1992.			

is a particularly accurate means of obtaining the initial portion of the time-strain curve and the constant rate of deformation of the secondary stage.

RESULTS

The present investigation was conducted principally with a cast 5 per cent chromium-molybdenum-silicon steel, the physical characteristics, heat treatment, and other pertinent information being shown in Table I, along with the results of the constant load, constant temperature tests. In Fig. 4, the secondary rate data is plotted by Kanter's (10) reciprocal absolute temperature scheme at various constant stresses. The lines thus determined are practically straight and parallel, which indicates that stress affects the energy index of creep but slightly. From the slope of these lines, a quantity of energy Q , which is independent of stress, may be calculated yielding a value

of 89,000 calories per gram mol. (The mol in this connection is the average crystal fragment or domain involved.) Correlation by this method extends over more than 5 log cycles or a ratio of rates of 100,000 to 1. The lower half of the figure covers the range of rates usually found in published creep data while the larger rates in the upper portion are obtained from rupture tests, usually of less than

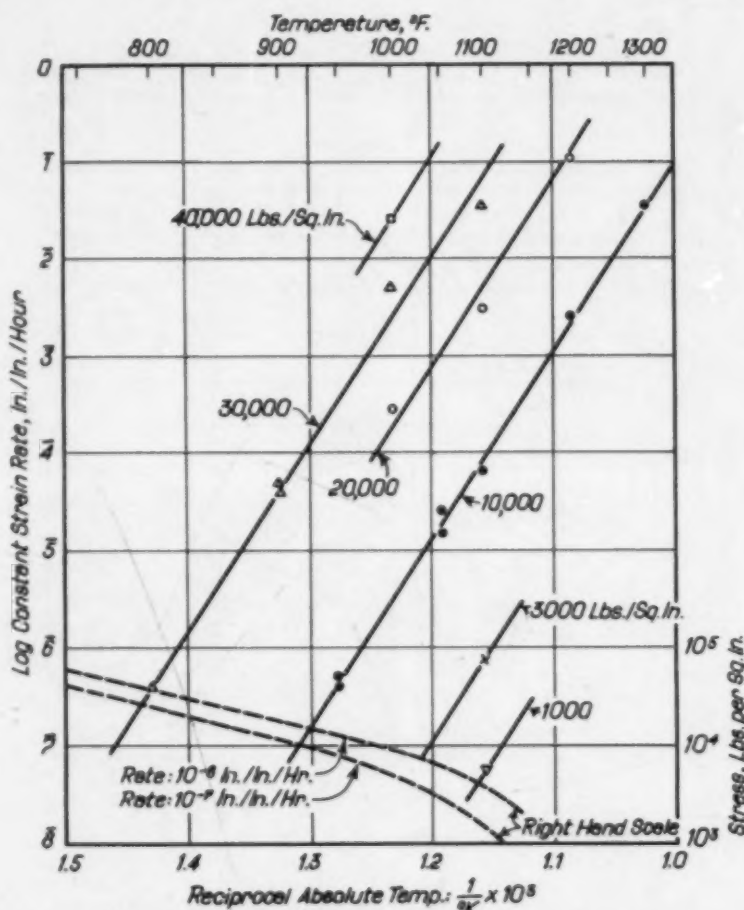


Fig. 4—Relationship Between Constant Strain Rate and Reciprocal Absolute Temperature at Constant Load for Cast 5 Per Cent Cr-Mo-Si Steel.

50 hours' duration. The line for 20,000 pounds stress was determined by rupture tests only and exhibits the proper slope and location.

A convenient means of interpolation for creep stress values is to draw lines of constant rate on a plot of reciprocal absolute temperature versus log of stress. Two of these lines are shown in the lower left-hand corner of Fig. 4. Thus, in this case, creep stress values for any temperature between 800 and 1100 degrees Fahr. may be obtained quite readily. It is noteworthy that points determined by the 20,000 pounds per square inch line, a product of short tests, agree

well with those determined by both the short and long tests upon which the 10,000 and 30,000 pounds per square inch lines are based.

The same secondary rate data is plotted in Fig. 5a, b, and c by the other methods mentioned previously. Only the actual test points are shown, others being obtained by interpolation from Fig. 4. While the lines may be very nearly straight over 2 or 3 log cycles of rates,

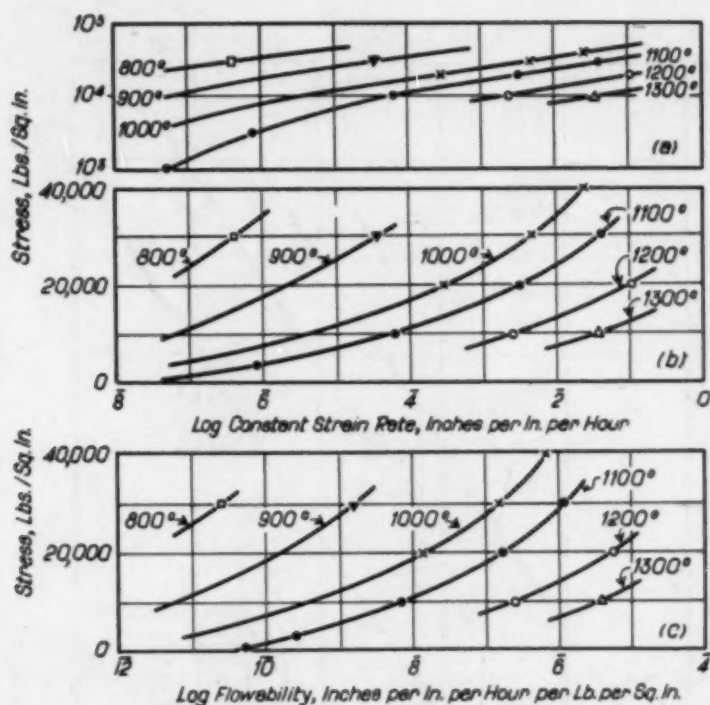


Fig. 5—Relationship Between Stress and Constant Strain Rate at Constant Temperature for Cast 5 Per Cent Cr-Mo-Si Steel.

- a. Log-Log Plot
- b. Semi-Log Plot
- c. Flowability Plot

in no case can a reliable extrapolation be made from short tests. Norton's plot in Fig. 5a shows that a straight line can be used to represent the data for higher stresses, in this case above 10,000 pounds per square inch, while it falls down at lower stresses. Ludwik's plot in Fig. 5b exhibits considerable curvature as does the flowability plot in Fig. 5c which was proposed for use only at the lower stresses where it has been demonstrated to be quite satisfactory.

Photomicrographs of the original structure and of the structures after some of the tests are shown in Figs. 6 to 9. It will be seen that the structures are very similar before and after test, apparently being unchanged by the conditions of the tests. This conclusion may not hold, however, for tests of longer duration at the same tempera-

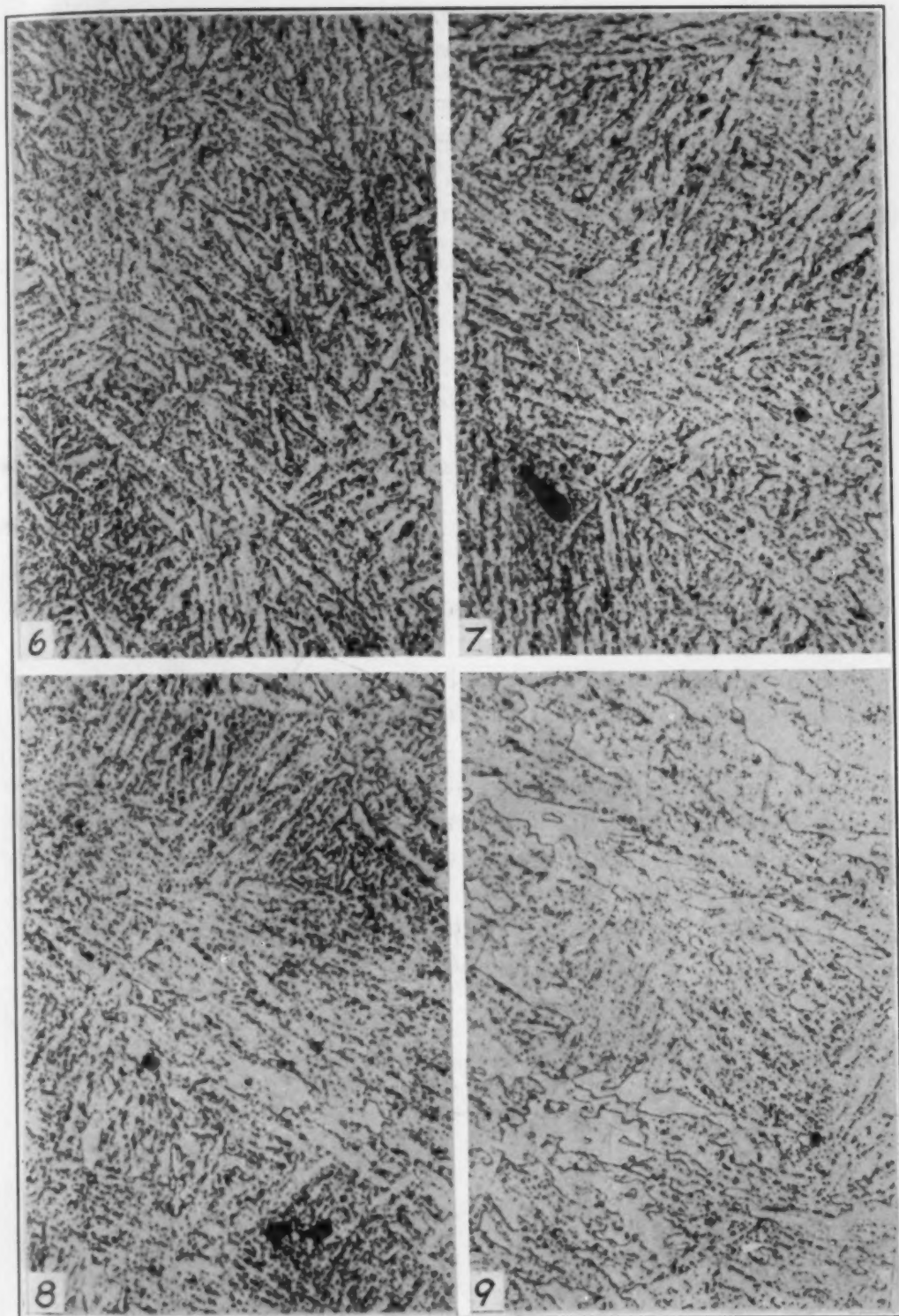


Fig. 6—Original Structure—5 Per Cent Chromium-Molybdenum-Silicon Steel. $\times 500$.
Fig. 7—Same After Rupture Test—1100 Degrees Fahr., 30,000 Pounds Per Square Inch, 1.98 Hours. $\times 500$.
Fig. 8—Same After Rupture Test—1200 Degrees Fahr., 20,000 Pounds Per Square Inch, 0.70 Hour. $\times 500$.
Fig. 9—Same After Rupture Test—1300 Degrees Fahr., 10,000 Pounds Per Square Inch, 7.1 Hours. $\times 500$.

Table II
Cast Carbon Steel

Analysis:	C	Si	Mn	S	P
	0.20	0.24	0.60	0.033	0.032
Heat Treatment:					
	Anneal 1550 degrees Fahr.				
Creep and Rupture Characteristics:					
Temp. °F.	Stress P.s.i.	Secondary Rate In./In./Hr.	Duration of Test Hours		
800	30,000	0.000102	360.		
900	30,000	0.0195	16.5		
800	20,000	0.0000101	528.		
900	20,000	0.000305	289.		
1000	20,000	0.023	14.6		
800	10,000	0.00000021	4012.		
850	10,000	0.00000112	384.		
1000	3,000	0.00000034	2848.		

tures. In judging this data, it should be remembered that more variability and nonuniformity are encountered in cast materials than in wrought, particularly with reference to grain size, which is known to exert such a potent influence over creep resistance.

A small amount of additional data was obtained on an annealed cast carbon steel and on a wrought copper-base alloy, the results appearing in Table II and Table III, respectively, along with other information. Fig. 10 presents a reciprocal absolute temperature plot of the secondary rates for both materials at various stresses. Once again, the steel yields an energy value of 88,000 calories per gram mol while the copper alloy, on which only short tests were run, gives 38,000 calories per gram mol. Both values are consistent with the slopes usually found in such analyses of the data for these materials.

Table III
Wrought Copper Alloy

Analysis:	Cu	Al	Si	Ni	Others
	90.13	7.00	2.03	Tr.	0.84
Heat Treatment:					
	Anneal 1300 degrees Fahr.				
Rupture Characteristics:					
Temp. °F.	Stress P.s.i.	Secondary Rate In./In./Hr.	Fracture Time Hours		
932	5000	0.0194	4.1		
1022	5000	0.074	1.22		
1112	5000	0.30	0.42		
1022	2500	0.0072	13.2		
1112	2500	0.030	4.0		
1112	1000	0.0040	50.		

CONCLUSIONS

It is believed that the method herein suggested has definite possibilities for using short-time test data to estimate creep rates by extrapolation. Successful use of this method depends primarily on one factor, namely, structural stability of the material over the entire temperature range of test and extrapolation at the stress used. A

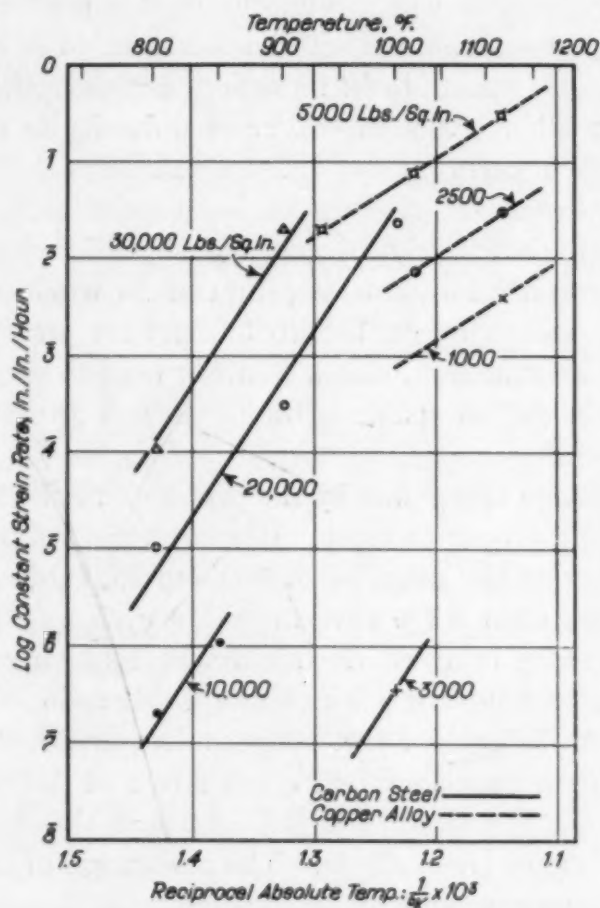


Fig. 10—Relationship Between Constant Strain Rate and Reciprocal Absolute Temperature at Constant Load for Cast Carbon Steel and a Wrought Copper Alloy.

structural change may lead to what appears like an abnormally high or low Q value, and while work in this vein has only begun, a few effects to be expected as contributing to inaccuracies might be mentioned. Spheroidization developing during a creep test results in an increasing creep rate, particularly at the higher temperatures. Application of a load in excess of the short-time yield point causes plastic deformation by slip and results in an extended primary stage

so that the "constant rate" portion may be only a transition stage between the primary and final stages instead of a true secondary or constant rate stage. Both of the previously mentioned factors would tend to yield apparently high Q values. Acting to give an apparently low energy value is the case of the relief of internal stresses as in cold-worked alloys tested at succeeding higher temperatures. Another structural change tending to produce a like effect is that of grain growth, a condition which is encountered in nonferrous alloys at moderately high temperatures. The separation of a new phase as in age hardening is uncertain in its action, but nevertheless a source of anomalous results. Another source of difficulty in employing the method would be in carrying tests on steels through the critical range.

There, furthermore, appears to be a possibility of detecting structural stability by correlating rate data by this scheme. If step-down tests are run in addition to step-up tests, it would seem that any disparity of Q values might be attributed to a structural change. Metallographic examination, which is often used to detect structural changes, is sometimes incapable of doing so and an indirect method could prove useful.

In extrapolating creep data by the proposed method, one is faced with establishing a proper Q value. However, the difference in using slope values covering the range of 85,000 and 95,000 found for steels is relatively small even when covering 6 log cycles. An uncertainty of 30 degrees Fahr. is about the maximum error involved for the temperature range explored. Since a 4 to 1 ratio of creep rates corresponds to a 30-degree Fahr. temperature spread, this is equivalent to a maximum uncertainty in creep stress of approximately 30 per cent. Usually a few short tests will determine the slope of the line well within the limits given above. The advantage of the reciprocal absolute temperature scheme of extrapolation is due largely to the fact that the slope of the constant stress line is found to be practically constant for a given material. Thus, knowing the slope, one test point determines the position of the line, although, of course, more points are desirable.

The authors wish to acknowledge their indebtedness to Crane Co. for permission to publish this information.

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DISCUSSION

Written Discussion: By E. R. Parker, Research Laboratory, General Electric Co., Schenectady, N. Y.

Messrs. Kanter and Sticha are to be complimented on this interesting exposition concerning the high temperature properties of metals and alloys. Although there is some scatter in the test data, the excellent agreement between short- and long-time tests is obvious.

This method of testing seems to have great promise for the rapid determination of the strength of metallurgically stable materials. However, it should be borne in mind that this test is a short-time test designed to evaluate materials for long-time use. Because it is a test which effectively condenses time, it should be the object of severe scrutiny and subjected to many independent checks.

If preliminary results can be considered as any sort of a check, some work now in progress in the Research Laboratory of the General Electric Company may be of interest. Some creep tests under constant load were conducted at several temperatures on single crystals of zinc. The activation energy for the creep of these crystals was of the same magnitude as that obtained for polycrystalline bars of the same material. Creep tests were also conducted on bars of polycrystalline lead and copper. The activation energy obtained for copper compares favorably with the value of 38,000 calories per mol obtained for the copper alloy described in the text.

The exponential relationship between the creep rate and the reciprocal of the absolute temperature seems to hold whenever temperature is the only variable.

Written Discussion: By B. B. Betty, Metallurgical Department, The International Nickel Co., Inc., Huntington, W. Va.

While this paper is indeed brief, as compared with some others with which Mr. Kanter's name is associated, its merit warrants for it a place among his other outstanding contributions to creep literature. The writer, too, has been interested for some time in a more rapid method of evaluating the relative merits of metals and alloys for high temperature service. The very precise methods of creep testing which have been sponsored and standardized by the American Society of Mechanical Engineers-American Society for Testing Materials Joint Research Committee on the Effect of Temperature on the Properties of Metals probably always will have a place in the field of metallurgical engineering. It must be recognized, however, that at present these methods are not well adapted to the several needs of the metal producers who must deal with a multiplicity of experimental alloys. For such work there is a great need for a rapid method of selecting the more desirable ones on the basis of their behavior at high temperatures.

The authors have put forward a new method of estimating secondary creep rates, in that they have shown that such information may be derived from rapid tests performed in a tension testing machine. In doing this they have rendered a real service to all those engaged in the metal producing industry. Furthermore, the value of the authors' paper lies almost as much in the fact that it breaks the shackles of tradition, as it does in the intrinsic worth of the knowledge that the method used for plotting data permits the bridging of a gap between a short-time creep test and the conventional long-time ones.

It is not expected that such a method of testing will completely relieve us of the necessity of making the present form of precise creep tests. It does provide, however, a safeguard against the overdevelopment of those expensive, tedious and time-consuming forms of experimentation.

The more publicity that is given to such developments and innovations in testing technique, the quicker will be their acceptance by the trade in general.

Written Discussion: By W. R. Burrows, engineering staff, Standard Oil Co., Whiting, Ind.

The authors are to be congratulated for this and other excellent contributions to our knowledge of the creep of metals.

This paper specifically attacks the problem of proper extrapolation of time-deformation creep characteristics obtained in relatively short periods of time. However, it also represents in our belief a further addition toward the establishment of analytical expressions truly representative of creep. Such expressions, when finally determined, will be of prime importance to the designer of high temperature equipment.

We note that for the most part laboratory studies consider creep properties of metals in simple tension. We feel that there is now a demand for the creep properties of metals subjected to two- or three-dimensional stress.

Simple tensile creep data have led to the use of design working stresses for

petroleum refining vessels based, say, on an observed tensile creep rate of 1.00 per cent in 10,000 hours. In using this method of design we have never heard of any creep resulting, or at least of any exceeding the probable error in the method of measurement. However, the statement of 1.00 per cent creep in 10,000 hours implies that a vessel 100 inches in diameter would increase about $8\frac{3}{4}$ inches in diameter in ten years of service. Thus there is a wide discrepancy between laboratory tensile creep results and observed creep due apparently to shear stress.

Of course, Nádaï, Bailey,¹ and others have given a good theoretical explanation of such discrepancies. They have demonstrated that creep in closed-end cylinders under hydrostatic pressure is due entirely to shear since no axial deformation is observed. However, their excellent theory is of little real practical value until experimental investigations establish a rational basis (in contradistinction to current wishful guessing) for determining the constants essential to its successful application. It may be of interest to note in passing that the independent theoretical investigations of Nádaï and Bailey lead to identical results.

We have been very much interested in reported simple tensile creep tests leading apparently to rupture after a relatively small amount of deformation. Here again we feel that two-dimensional creep in cylinders may be radically different from simple tensile creep. Observed creep failures in carbon steel refinery furnace tubes indicate that possibly as much as 100 per cent local creep occurs before the longitudinal split appears in the bulged portion. A second example is furnished by our arbitrary practice of rejection of refinery furnace tubes when their external circumference indicates 5.00 per cent creep. Since those tubes that we have observed are heated almost entirely on one side, this 5.00 per cent must necessarily represent an average value. The instantaneous maximum is probably, therefore, greatly in excess of 5.00 per cent. Nevertheless, no signs of failure due to creep are observed.

In this discussion of the author's work we have digressed considerably from the actual subject matter at hand in the hope that future work in connection with extrapolation of creep data and establishment of analytical expressions will progress along lines of practical interest to the designer.

Written Discussion: By T. McLean Jasper, A. O. Smith Corp., Milwaukee.

The writer believes that tests of short duration at elevated temperatures are not likely to produce reliable data for steel to operate at high temperatures for long periods of time. The fundamental reason is that a single chemical composition of steel is offered for use in a great variety of states of treatment to meet the same specification for pressure vessels, and depending upon the temperature of operation these states of treatment will persist for a longer or shorter period of time. I would refer to *Proceedings of the American Society for Testing Materials*, Vol. 28, 1928, p. 99, Fig. 15, to bear out my contention. The authors might suggest that the steel be tested after annealing has been performed because by the figure referred to above it is shown that the creep

¹A. Nádaï, "Plasticity", 1931, p. 203, and R. W. Bailey, "Thick-Walled Tubes and Cylinders Under High Pressure and Temperature", *Engineering* (London), June 20, 1930, p. 786.

curves for annealed, normalized and quenched and drawn 0.35 carbon steel approach the same slope after about 350 hours for the temperature of testing which was about 800 degrees Fahr. The annealed steel reaches a normal slope much earlier than the other states. For high temperatures of operation, this might possibly be a helpful suggestion.

The machine and pressure vessel designing engineer is at present in a "jittery" state of mind with reference to the application of creep test data which has shown him results which vary widely for the same steel and applied to types of steel he is in the habit of using most. Then the creep values he is given do not appeal to his engineering sense of a factor of safety based on a failure or breakdown of his design. It is believed that the long time ultimate strength values and the long time yield values for steel at high temperatures do supply him with data which appeal to his experience so that his engineering judgment may also be employed in designs covering various degrees of hazard.

It must be recognized that the fundamental deflection equations applied to steam turbines, for instance, do not depend on creep data as much as on stiffness factors involving the section modulus and the moduli of elasticity in shear and direct stress. These values for steel at the temperature of operation are fundamentally necessary.

The writer does not desire to discuss the authors' method of handling creep data. The point desired to be made is the value of creep tests as such influenced as they are by so many factors such as initial internal stresses, corrosion during testing, and difficulties with the measuring and constancy of the heat control equipment. Then, too, values are given as stress causing 1.00 per cent stretch in 100,000 hours, or 1.00 per cent in 10,000 hours, and these vary widely. Which should the engineer use and why? Why not 1.00 per cent in 65,000 hours, or 1.00 per cent in 2000 hours? The engineer certainly desires to be able to use some judgment based on the hazards associated with his designs.

Oral Discussion

H. F. MOORE:² The authors of this paper have pushed the important problem of a short-time test which will be a fairly reliable index of creep-resisting properties of metals to a point where the problem seems *possible* of solution, at least for some metals. However, all short-time tests when used to predict long-time service necessarily involve extrapolation. We do not like to use extrapolation of data, but again and again we are compelled to do so,—at least in the world of applied science. Extrapolation may be likened to the small boy's definition of a lie, "An abomination unto the Lord and a very present help in trouble".

The authors have worked out a scheme whereby the diagrams used are straight lines. I cannot see any greater justification for the extrapolation of a straight line than for the extrapolation of a curved line, but I will admit that the extrapolation of a straight line is more convenient, and more psychologically satisfactory. In passing, may I note that extrapolation from a small quantity to zero may be just as dangerous as extrapolation from a large quantity to infinity. If you do not believe this, plot values on logarithmic paper.

²Research professor of engineering materials, University of Illinois, Urbana, Ill.

The authors' statement as to the necessity of stability of the material is worthy of emphasis. In the Investigation of Creep and Fracture of Lead Sheathing, now in progress at the University of Illinois, it has so far been found impossible to obtain Mr. Kanter's straight-line diagrams. Our tests were run at 110 degrees Fahr., at which temperature some lead alloys are probably unstable as crystalline structures. This instability, or possible atmospheric corrosion, may explain this inability to apply Mr. Kanter's results to lead and lead alloys.

Authors' Reply

We would like to briefly express our thanks for the discussions which have been offered and at the same time point out that there seems to be some little disregard, as there usually is in discussions, for the particular point that is at issue. We are by no manner of means advocating the abandonment of the long-time creep test to get fundamental design information. To bear that out, it might be stated that we favor increased facilities for making long-time creep tests. But, at the same time, there is a very crying need for a short-time inspection method or acceptance method to use in determining whether a given lot of material will fulfill the particular design requirements for which long-time tests have established its desirability, and for those types of materials which have been empirically proven to be stable and well behaved. The type of extrapolation that we are advocating may be useful as the basis for some kind of an arbitrary acceptance test. To bear this out, it might be added that the method, in some further work which is now in progress on a steel in two different conditions of heat treatment, reflects those differences which the heat treatments establish for creep properties.

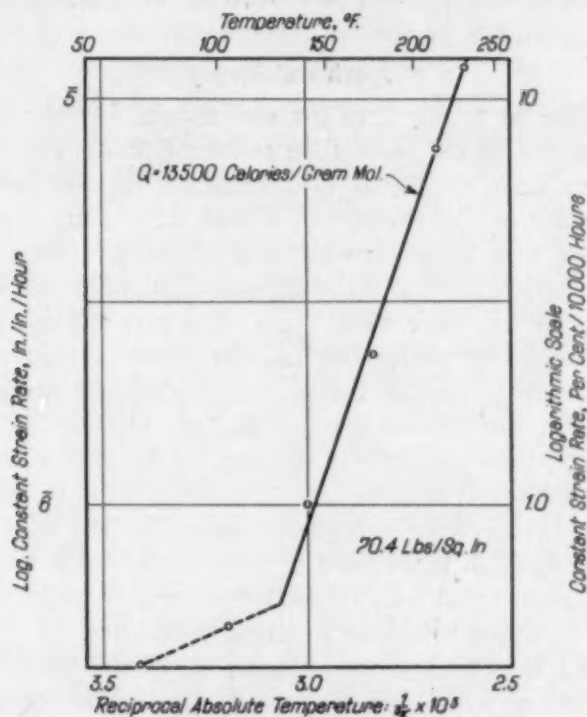
We have tried to refrain from a discussion of the theoretical implications of the type of plotting which Mr. Parker touched upon because for our present purpose we are only interested in showing how the reciprocal absolute temperature plot might be utilized for shortening the creep testing necessary on certain steels in the range of temperature in which they are being used.

The possibility of using this method of testing for the rapid determination of the strength of metallurgically stable materials has been very nicely strengthened by Mr. Parker's remarks. The need for stability cannot be emphasized too strongly. While the effect of a structural change on the observed rate in a rupture test may be small, this same change may exert a very pronounced effect on the rate as determined in the usual creep test even after 1000 hours of testing. Failure to get a straight line correlation or a proper slope value can often be traced to a condition of this sort. Mr. Parker has voiced our own thoughts in suggesting severe scrutiny and independent checks of the proposed method.

The information relative to the creep of single crystals of zinc as compared with polycrystalline zinc is most interesting, and we are looking forward to the time when this data and that on lead and copper are published.

In applying the proposed method to new alloys, Mr. Betty had best proceed with caution. Lack of structural stability, as produced by heat treatment or otherwise, or improper choice of rupture test conditions may result in the rejection of a desirable alloy.

Professor Moore's inability to obtain straight line diagrams may be explained, as he himself points out, by instability or recrystallization in the temperature range under investigation. The results of Hanffstengel and Hanneman ("Mechanism of Creep and Fatigue Limit Investigated on Lead and Lead Alloys," *Zeitschrift Metallkunde*, 1937, Vol. 29, No. 2, p. 50-52) at somewhat higher temperatures definitely above the recrystallization range yield a straight



Reciprocal Absolute Temperature Plot of Hanffstengel and Hanneman Creep Data for Soft Lead Under a Stress of 70.4 Pounds Per Square Inch.

line, but at lower temperatures there is considerable departure. It is quite possible that the explanation of this departure is not as simple as indicated above. A difference in the mechanism of creep is suggested as another possibility.

The remarks of Mr. Burrows and Mr. Jasper, while of interest, are essentially design problems dealing with the application of creep test data, and as such are outside the scope of this paper.

THE STRESS-STRAIN CHARACTERISTICS OF THE TORSION IMPACT TEST

BY O. V. GREENE AND R. D. STOUT

Abstract

This paper presents a method of analyzing the torsion impact test for stress-strain characteristics. The stress-strain properties of hardened and drawn 1.10 per cent carbon tool steel in torsion are calculated from subrupture impact tests and static tests. Both methods show a good agreement.

A clear conception of the strength and ductility factors of the torsion impact test is provided by the stress-strain curves.

Elastic limits obtained from the static test are reported for the complete drawing range.

A rapid method of determining deformation and average stress values of rupture is described.

An appendix is attached containing a full derivation of the mathematical formulas necessary for evaluating the stress-strain curves.

INTRODUCTION

THE interest existing in the field of impact testing is reflected by numerous recent publications dealing with this subject. In the last few years, attempts have been made to reduce the variables involved in impact testing to certain fundamental factors. The data obtained from an impact test is usually expressed as work, which is a direct measure of the toughness of the material. It is a common misconception to consider toughness as synonymous with ductility. Toughness is composed of two equally important components, which are strength and ductility. These terms correspond respectively to the force and the distance components of the work necessary to cause rupture. This conception is well explained by Jeffries and Archer (1).¹ As in other impact tests, torsion impact values are a measure of the toughness of the material in foot-pounds, but do not indicate the relative magnitude of the strength and plasticity char-

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, O. V. Greene is associated with the metallurgical department, The Carpenter Steel Company, Reading, Pa., and R. D. Stout is instructor in metallurgy, Lehigh University, Bethlehem, Pa. Manuscript received June 20, 1939.

acteristics of the specimen. Consequently, it was considered desirable to investigate the relative importance of these factors of torsion impact values and to measure their actual magnitude in 1.10 per cent carbon tool steel after various heat treatments.

The torsion impact test has been described in detail, and the torsion impact properties of 1.10 per cent carbon tool steel have been reported in previous papers (2), (3), (4). The torsion test has been also used by Emmons (5), (6), (7) to measure the toughness of hard, low ductility materials, but under conditions of static rather than impact loading.

METHODS OF INVESTIGATION

In the analysis of the torsion impact test, there are several methods by which stress and strain data can be evaluated. Since the unit stress developed at any point in a section under torsion is dependent on the distance of that point from the neutral axis, and thus varies from zero at the center to a maximum at the surface, it is impossible to measure stress directly, but instead it must be calculated from torque values. Incidentally it should be noted that the area under torque-strain curves represents the work required to cause rupture. On the other hand, deformation can be measured directly on the specimen. However, in the fraction of a second involved in the torsion impact test, torque-strain measurements are impossible. The plastic deformation or strain can be estimated by scribing a line on the gage length of the specimen before testing, and refitting the pieces after testing to note the twist. This is, of course, difficult on specimens that have not been drawn sufficiently high to cause failure by shear rather than by shattering. Shattering usually occurs on torsion impact tests of specimens of 1.10 per cent carbon tool steel when drawn below 350 degrees Fahr. Two methods which were used to supply torque-strain data in this investigation are as follows:

The Torsion Impact Test—It is possible to use a number of torsion impact specimens in tests in which the rotation of the striking member is limited to a series of speeds sufficient to cause twist of a specimen but not high enough to result in rupture. Thus a relation between the degrees of twist or ductility and the work in foot-pounds required to produce the twist will be obtained, from which a torque-strain curve can be constructed. The method used in accomplishing this will be explained more fully.

The Static Torsion Test—In this method of testing a specimen can be loaded slowly and the deformation noted for each increment. This produces a torque-strain curve directly.

From the torque-strain curves obtained by either method, stress-strain curves can be derived.

Previous studies (8) of the effect of speed on impact and static testing have shown marked discrepancies between fast and slow

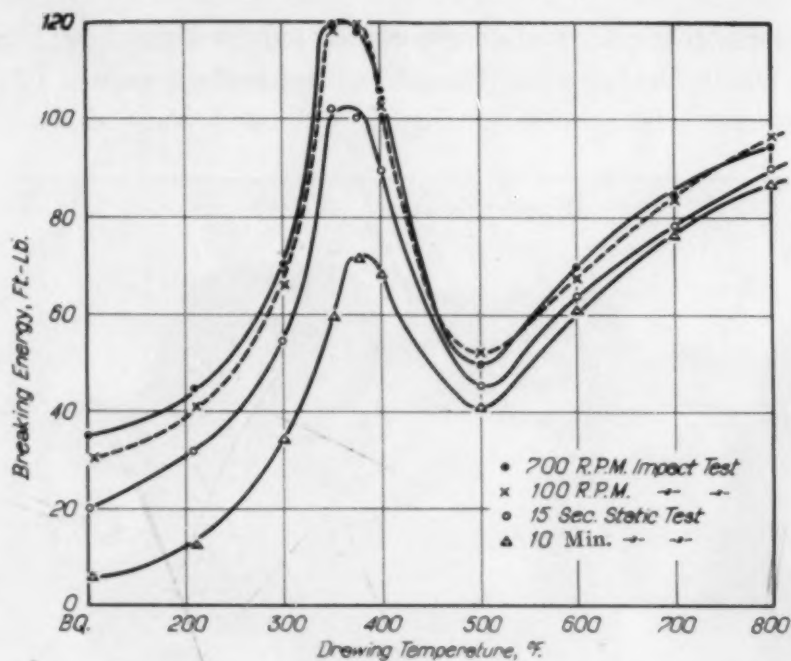


Fig. 1—Effect of Drawing Temperature on Torsion Values of 1.10 Per Cent Carbon Steel Tested at Various Speeds. Hardened by Brine Quench from 1450 Degrees Fahr. and Drawn as Shown.

testing. The relations between resistance to plastic flow and cohesive strength have been studied to explain these effects and to express these characteristics in forms of laws. It has also been found that the nature of the material has a marked influence in this respect. In the present case the effect of speed over a range of 700 revolutions per minute (73 radians per second) in the torsion impact test at one extremity, to a static test lasting ten minutes (0.001 radians per second) at the other, has been observed, and reported by the authors (9). Fig. 1 shows a series of curves summarizing this study. These indicate that there is sufficient agreement between impact and 15-second static tests to justify the application of data from static tests and low speed subrupture impact tests to the analysis of torsion impact characteristics. Incidentally this effect of

speed on test results has no connection with the higher stresses introduced by dynamic loading as compared to static loading, i.e., the action of live versus dead loads, where the stress increase is due entirely to inertia and to the forces of deceleration rather than the changes in the behavior of the loaded member itself.

TEST PROCEDURE

The torsion impact tests were conducted on a machine as shown in Fig. 2, which has been fully described in previous papers (2), (3). An arrangement illustrated in Fig. 3 was used for the static tests.

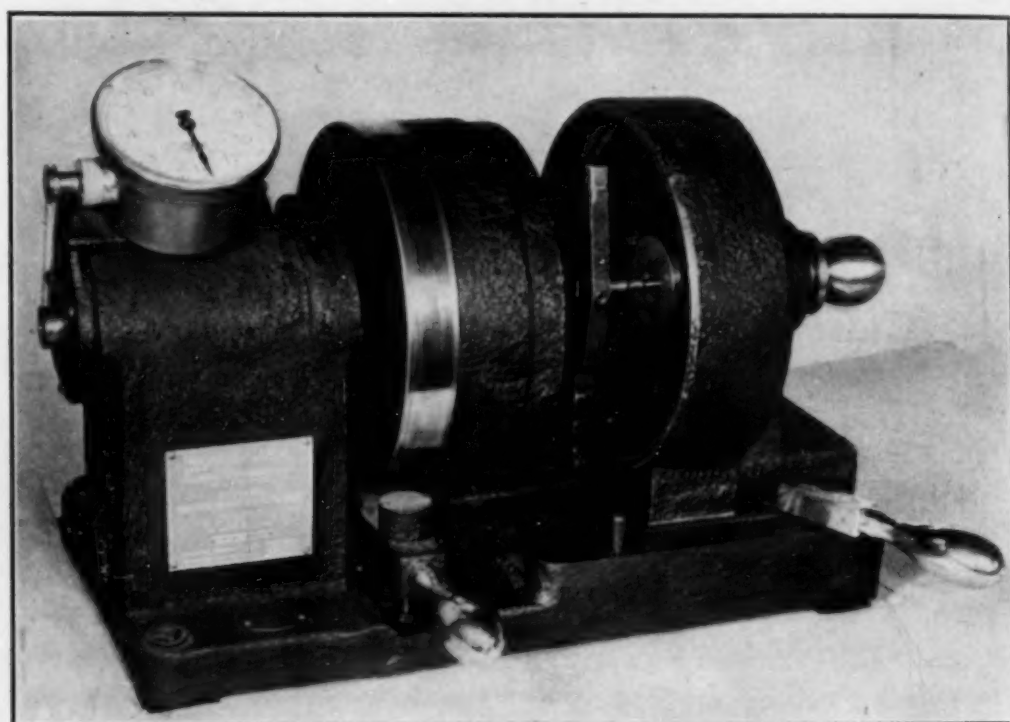


Fig. 2—Torsion Impact Machine.

Torque was applied by pulley "A" and a weight "B" using sand as the loading agent. Deformations were read by means of a clinometer "F" and "G" fastened on the ends of the specimen "C". The specimen is gripped by chucks "D" and "E".

A heat of carbon tool steel of the following analysis was used throughout this investigation:

Carbon Per Cent	Manganese Per Cent	Silicon Per Cent	Phosphorus Per Cent	Sulphur Per Cent
1.10	0.25	0.27	0.018	0.016

Specimens were hardened by quenching in brine after heating for 3 minutes in a lead bath at 1450 degrees Fahr. (790 degrees Cent.). Specimens were then drawn at 212, 300, 350, 400, 500, 600, 700 and 800 degrees Fahr. The standard torsion impact specimen shown in Fig. 4 was used both for the static and the impact testing. Specimens were treated to permit static and impact tests in quadruplicate and also a series of torsion impact tests below rupture for each drawing temperature. The static tests were made in 15 seconds in order to approach impact conditions as closely as possible. This is especially

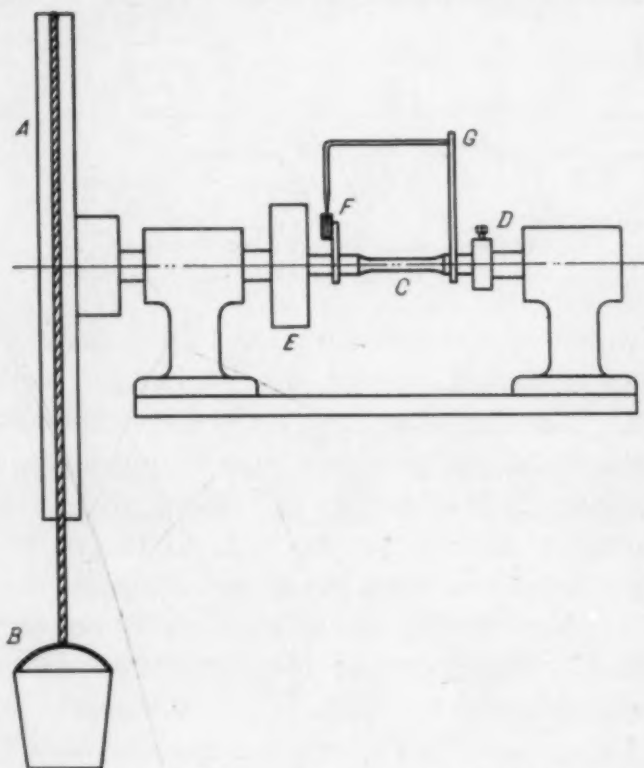


Fig. 3—Diagrammatic Sketch of Torsion Impact Machine.

necessary in the lower drawing temperature range. The reason for choosing a rapid static test is obvious by referring again to the trends shown in Fig. 1.

Impact Tests Below Rupture—In testing specimens below rupture, the rotor of the torsion impact machine is operated at a succession of speeds too low to cause failure of the specimen, but sufficient to effect varying amounts of plastic deformation. In such a test, this plastic deformation results in a certain amount of permanent set, and the specimen springs back elastically, stalling and reversing the rotation of the flywheel. This elastic recovery necessitated

the use of a cross arm with hinged ends to prevent the tups of the rotor from striking the specimen a second time but from the opposite direction. The specimens after testing below rupture were removed from the machine and the permanent deformation in degrees measured by a specially constructed protractor. Consequently the relation between the degrees of twist and the work required for this amount of deformation was obtained.

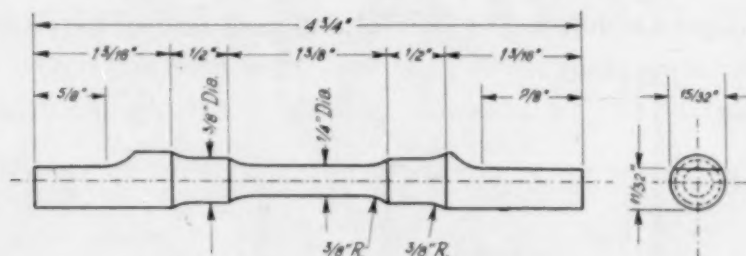


Fig. 4—Torsion Impact Test Specimen.

The derivation of stress-strain curves from these data is developed in the appendix. However, the method of analysis will be described here. The first step involves the conversion of the work and the permanent deformation data to a torque-strain curve. This is accomplished as shown in Fig. 5. Below the elastic limit, the torque developed is directly proportional to the deformation in a relation which is dependent upon the shear or torsion modulus. This is shown in Fig. 5 as the "modulus line." The plastic deformation can be measured by the departure from this line. If a specimen is twisted both elastically and plastically below rupture, it will spring back along a line parallel to the original modulus line but displaced from it by an amount equal to the permanent set. It is obvious then that the maximum torque and total deformation values of the specimen must lie somewhere along this line. Hence it is possible to erect a series of lines each representing a subrupture test. In Fig. 5 these lines are shown for a specimen brine quenched at 1450 degrees Fahr. (790 degrees Cent.) and drawn at 350 degrees Fahr. (175 degrees Cent.). Each line is marked with the foot-pounds of energy used for the test, and the distance it lies from the modulus line indicates the resultant permanent twist.

It follows that the area under the torque-strain curve of each of these test values must correspond to the energy used in each test. The elastic limit can be used for the first point on the curve. In

order to determine the elastic limit, a specimen is tested at successively higher speeds until the first measurable set appears. From the energy value of the elastic limit in foot-pounds the torque and deformation can be calculated and plotted in the diagram. It is known that the area under the curve from the elastic limit to a point on the first parallel line, which was the 10-foot-pound line in

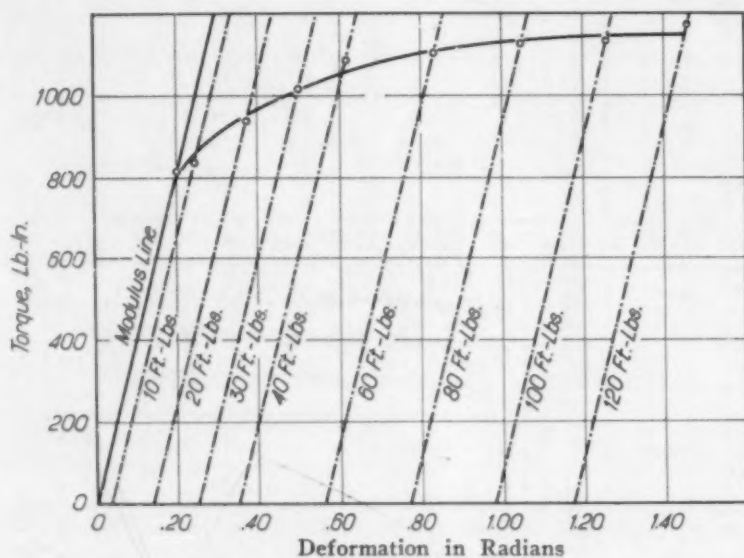


Fig. 5—Calculated Torque-Strain Curve from Impact Tests Below Rupture. 1.10 Per Cent Carbon Tool Steel Brine Quenched from 1450 Degrees Fahr. and Drawn at 350 Degrees Fahr.

this case, must be equal to the difference in the energies involved. By an algebraic solution shown in the appendix, the torque and total deformation values are calculated respectively for this line and plotted. In this fashion a full set of points for the torque-strain curve can be obtained.

The calculation of stress values from the torque-strain curve is accomplished by means of Equation 1 which is derived in the appendix, and is an adaptation of a similar equation developed by Foepl (10). This equation is as follows:

$$S = \frac{1}{2\pi r^3} \left(\frac{dT}{d\Phi} \Phi + 3T \right) \quad \text{EQUATION (1).}$$

Where

S = maximum fibre stress in pounds per square inch

r = specimen radius

T = torque in pound inches

Φ = deformation in radians.

$\frac{dT}{d\Phi}$

— = This expression represents the slope of the curve at the

$d\Phi$

particular point selected. By application of the formula to a series of points, values for a stress-strain curve are secured.

In Fig. 6 are given the stress-strain curves for the complete drawing range. Values for the co-ordinates of these curves are contained in Table I.

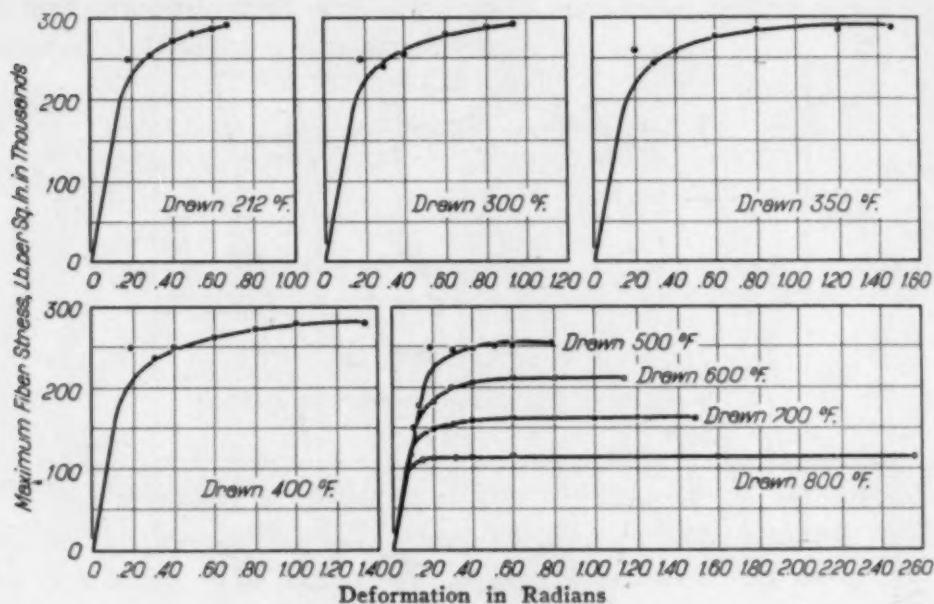


Fig. 6—Impact Tests Below Rupture. Stress-Strain Characteristics in Torsion of 1.10 Per Cent Carbon Tool Steel Brine Quenched from 1450 Degrees Fahr. and Drawn as Shown.

Static Tests—The average values obtained from four static tests were plotted as torque-strain curves for each drawing temperature. In Fig. 7 is shown a curve representing specimens drawn at 350 degrees Fahr. The straight portion of the curve represents the deformation below the elastic limit. The departure of this curve from this line denotes the plastic deformation. The area under the curve represents the work required to rupture the specimen. From this curve it is possible to obtain elastic limit, maximum torque, total deformation, and by integration, a work value. In Table II these values are given for each drawing temperature, together with torsion impact work values and Rockwell C hardnesses. The degree of agreement between static and impact tests is again indicated.

The values of maximum fiber stress existing at the surface of the specimen which are required for the stress-strain relationship can be calculated directly from the torque-strain curve by means of Equation 1. The calculated static stress-strain curve is shown for each drawing temperature in Fig. 8. The data used for the con-

Table I
Data for Stress-Strain Curves Calculated from Torsion Impact Tests Below Rupture

Maximum Fiber Stress—Pounds Per Square Inch
Strain—Radians.

1.10 Per Cent Carbon Tool Steel, Brine Quenched 1450 Degrees Fahr., Drawn as Shown:

Strain	Stress	Strain	Stress
Drawn at 212 Degrees Fahr. Rockwell C-66/67		Drawn at 500 Degrees Fahr. Rockwell C-58/59	
0.19	249,000	0.19	249,000
0.30	255,000	0.30	246,000
0.40	270,000	0.40	247,000
0.50	282,000	0.50	250,000
0.60	288,000	0.60	251,000
0.66	292,000	0.70	252,000
Drawn at 300 Degrees Fahr. Rockwell C-66/67		Drawn at 600 Degrees Fahr. Rockwell C-54/55	
0.19	249,000	0.13	176,000
0.30	239,000	0.20	182,000
0.40	254,000	0.30	199,000
0.60	280,000	0.40	209,000
0.80	288,000	0.60	210,000
0.94	293,000	0.80	210,000
Drawn at 350 Degrees Fahr. Rockwell C-63/64		Drawn at 700 Degrees Fahr. Rockwell C-51/52	
0.20	260,000	0.11	150,000
0.30	242,000	0.20	148,000
0.40	257,000	0.30	156,000
0.60	276,000	0.40	162,000
0.80	284,000	0.60	163,000
1.20	285,000	0.80	163,000
1.46	286,000	1.00	164,000
Drawn at 400 Degrees Fahr. Rockwell C-61/62		Drawn at 800 Degrees Fahr. Rockwell C-48/49	
0.19	249,000	0.09	120,000
0.30	235,000	0.15	110,000
0.40	248,000	0.20	113,000
0.60	262,000	0.30	114,000
0.80	274,000	0.40	114,000
1.00	279,000	0.60	115,000
1.34	280,000	2.56	115,000

struction of these curves are given in Table III. As in the impact stress-strain curves, the variation of stress levels and deformation with drawing temperature is evident from the height and length respectively of this series of curves.

DISCUSSION OF RESULTS

The derivation of the stress-strain characteristics of torsion impact values as shown in Fig. 6 gives a clear conception of the strength and ductility factors of the torsion impact test. The stress closely follows the Rockwell C hardness over the drawing range, while the strain increases and decreases similarly to the impact values. However, the interdependence of strength and ductility in controlling toughness is illustrated by a comparison between the

curves of the draws at 350 and 800 degrees Fahr. The test at 800 degrees Fahr. shows nearly twice the ductility but less than half the strength of the test at 350 degrees Fahr. Consequently the toughness values of the two drawing temperatures approach one another, with that of 800 degrees Fahr. slightly lower, as shown in Fig. 1 and Table II.

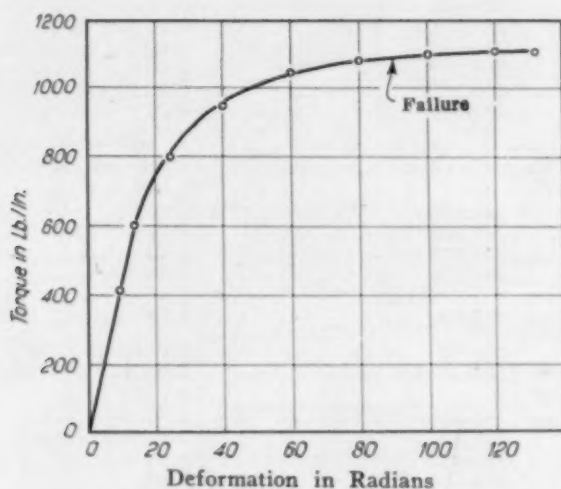


Fig. 7—Typical Torque-Strain Curve of 1.10 Per Cent Carbon Tool Steel Brine Quenched from 1450 Degrees Fahr. Drawn at 350 Degrees Fahr. Static Test Run in 15 Seconds.

These curves also indicate that the drop in torsion impact values in the range between 475 and 525 degrees Fahr. is due almost entirely to a loss of ductility, since the strength is only slightly lessened. This loss of ductility has been demonstrated previously in a paper (4) in which specimens were scribed to permit measurement of the twist.

It is interesting to note the very high stress levels exhibited by 1.10 per cent carbon steel tested in torsion. The method of detecting the elastic limit by impact necessitates an appreciable set in the specimen and thus leads to somewhat high values. This can be seen in Fig. 6 where the first points lie above the curves.

The stress-strain curves in Fig. 8 derived from the static tests seem to be quite similar to the subrupture tests in Fig. 6. This similarity is more conveniently shown in Fig. 9 for four of the drawing temperatures. In this figure a direct comparison is made between 15-second static and subrupture impact tests. The somewhat higher stresses calculated from the subrupture tests are probably an effect of the speed of testing. However, at low drawing temperatures, the

Table II
Data from 15-Second Static Torsion Tests

1.10 Per Cent Carbon Tool Steel, Brine Quenched 1450 Degrees Fahr. and Drawn as Shown:						
Drawing Temperature	Maximum Torque Lb.-Ins.	Total Deformation Radians	Integrated Work Ft.-Lbs.	Torsion Impact Values Ft.-Lbs.	Rockwell "C"	Elastic Limit Lbs. Per Sq. In.
As quenched	980	0.48	20	35	66/67	170,000
212° F.	1020	0.60	32	45	66/67	175,000
300° F.	1130	0.89	54	70	65	175,000
350° F.	1120	1.31	102	120	63/64	170,000
400° F.	1060	1.22	90	105	61/62	170,000
500° F.	910	0.81	45	50	58/59	160,000
600° F.	790	1.14	64	70	54/55	115,000
700° F.	650	1.59	78	85	51/52	100,000
800° F.	480	2.54	90	95	48	90,000

lower indicated work values of static testing as shown in Table II are largely due to lower deformations rather than lower stress levels. This is also an effect of speed, which has been discussed in a recent paper by the authors (9). These differences of stress and deforma-

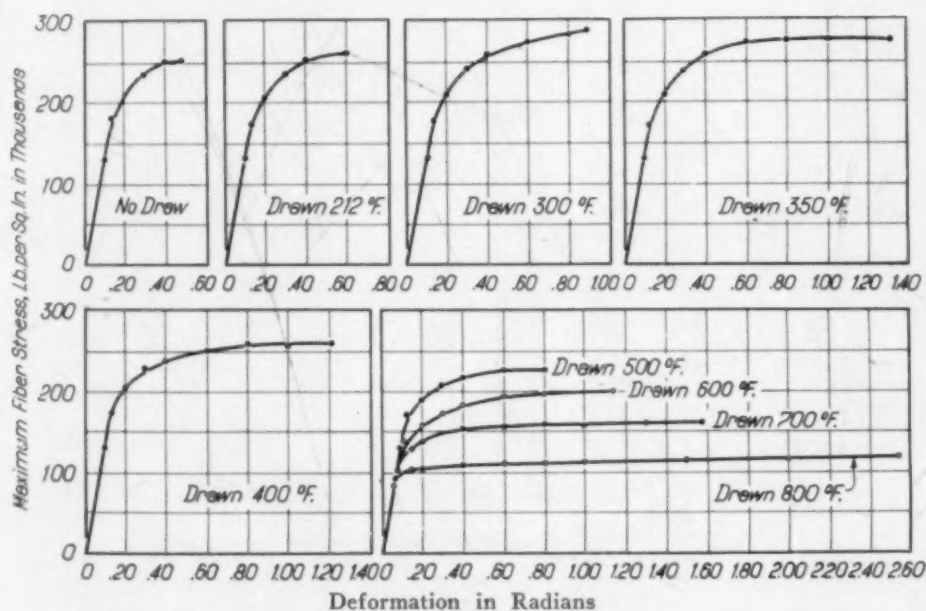


Fig. 8—Stress-Strain Characteristics in Torsion of 1.10 Per Cent Carbon Tool Steel Brine Quenched from 1450 Degrees Fahr. and Drawn as Shown. 15 Sec. Static Test.

tion are not sufficient to necessitate the division of these two methods of test into separate classes.

Elastic limits are readily determined in the static test and are given in Table II. Throughout the drawing range they are all approximately 60 per cent of the breaking strength.

It is evident that the complete determination of the stress-strain

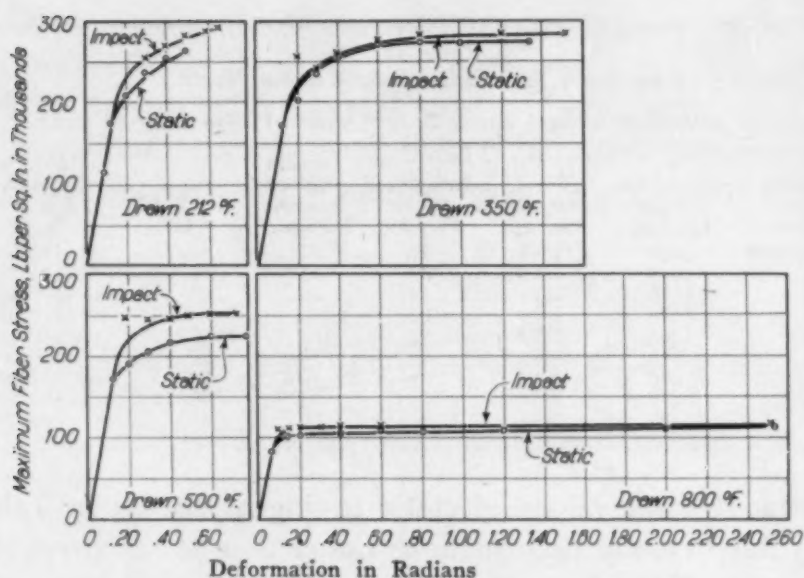


Fig. 9—Comparison of Stress-Strain Curves Obtained by Static Tests and Impact Tests Below Rupture of 1.10 Per Cent Carbon Tool Steel Brine Quenched from 1450 Degrees Fahr. Drawn as Shown.

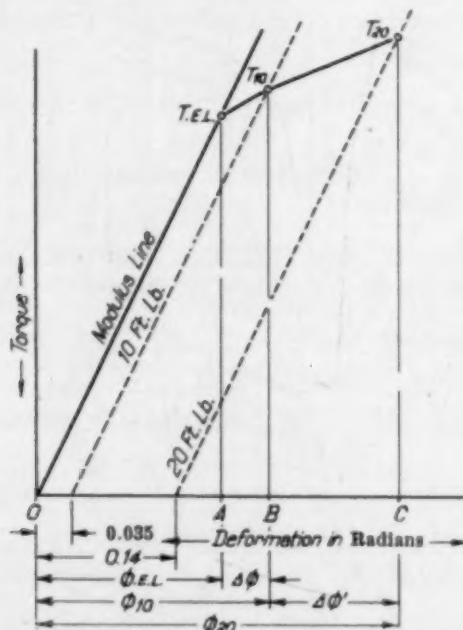


Fig. 10—Method of Construction of Torque-Strain Curve from Impact Tests Below Rupture.

characteristics of a material from torsion impact data is somewhat laborious. Hence, since the torsion impact test more nearly approaches service conditions than does the static test, it would be desirable to have a simpler method of calculation available. For this reason an approximate method for stress-strain evaluation will

Table III
Data for Stress-Strain Curves Calculated from Static Torsion Tests

Maximum Fiber Stress—Lbs. Per Sq. In.			
Strain-Radians			
1.10 Per Cent Carbon Tool Steel, Brine Quenched 1450 Degrees Fahr., Drawn as Shown:			
Strain	Stress	Strain	Stress
As quenched		Drawn at 500 Degrees Fahr.	
Rockwell C-66/67		Rockwell C-58/59	
0.10	131,000	0.13	170,000
0.14	180,000	0.20	189,000
0.20	207,000	0.30	208,000
0.30	236,000	0.40	217,000
0.40	252,000	0.60	226,000
0.48	253,000	0.81	226,000
Drawn at 212 Degrees Fahr.		Drawn at 600 Degrees Fahr.	
Rockwell C-66/67		Rockwell C-54/55	
0.10	131,000	0.10	137,000
0.13	173,000	0.20	158,000
0.20	206,000	0.30	172,000
0.30	236,000	0.40	184,000
0.40	253,000	0.60	193,000
0.60	261,000	0.80	196,000
		1.14	197,000
Drawn at 300 Degrees Fahr.		Drawn at 700 Degrees Fahr.	
Rockwell C-66/67		Rockwell C-51/52	
0.10	131,000	0.06	90,000
0.13	176,000	0.15	130,000
0.20	208,000	0.20	138,000
0.30	240,000	0.40	152,000
0.40	258,000	0.60	156,000
0.60	274,000	0.80	157,000
0.80	284,000	1.00	157,000
0.89	288,000	1.30	160,000
		1.60	161,000
Drawn at 350 Degrees Fahr.		Drawn at 800 Degrees Fahr.	
Rockwell C-63/64		Rockwell C-48/49	
0.10	131,000	0.06	85,000
0.13	173,000	0.15	103,000
0.20	206,000	0.20	104,000
0.30	239,000	0.40	108,000
0.40	258,000	0.60	109,000
0.60	274,000	0.80	109,000
0.80	275,000	1.00	111,000
1.00	275,000	1.50	113,000
1.31	275,000	2.00	116,000
		2.54	118,000
Drawn at 400 Degrees Fahr.			
Rockwell C-61/62			
0.10	131,000		
0.13	173,000		
0.20	204,000		
0.30	230,000		
0.40	238,000		
0.60	252,000		
0.80	258,000		
1.00	259,000		
1.22	262,000		

be described. The only data required are total work and permanent deformation. In order to measure permanent deformation, the specimen can be scribed before testing, and after the test the twist of the refitted pieces can be determined. The work value of the rupture in foot-pounds and the deformation in degrees twist can be applied to an equation derived previously (4). This equation can

Table IV
Comparison of Maximum Fiber Stress at Rupture Calculated by Various Methods

1.10 Per Cent Carbon Drawing Temperature Degrees Fahr.	Stress—Pounds Per Square Inch Tool Steel, Brine Quenched 1450 Degrees Fahr., Drawn as Shown:		
	Average Stress Equation Using Torsion Impact Data	Torsion Impact Test Below Rupture	Static Torsion Test
212	272,000	292,000	261,000
300	248,000	293,000	288,000
350	250,000	286,000	275,000
400	240,000	280,000	262,000
500	221,000	252,000	226,000
600	186,000	210,000	197,000
700	160,000	164,000	161,000
800	105,000	115,000	118,000

be used to evaluate the stress in the section at the point of rupture. Equation is:

$$S = 10,000 (\sqrt{2.9 a^2 + 54W} - 1.71 a)$$

Where

S = The average stress

W = Torsion impact in foot-lbs.

a = Permanent deformation in degrees.

This equation has been applied to torsion impact data and the results as compared to values from the static and subrupture impact tests are contained in Table IV.

SUMMARY

1. Torsion impact test can be analyzed by appropriate means and the stress-strain characteristics derived.
2. The stress-strain properties of 1.10 per cent carbon tool steel have been calculated by two methods, and it has been shown that there is a good agreement in values obtained from both static and impact tests.
3. A clear conception of the strength and ductility factors in the torsion impact test is provided by the stress-strain curves.
4. Elastic limits in torsion have been determined and found to be approximately 60 per cent of the breaking strength throughout the drawing range employed in this work.
5. A rapid method of obtaining average stress and strain values at rupture is described.

ACKNOWLEDGMENT

The authors wish to acknowledge their indebtedness to Mr. G. V. Luerssen for his valuable suggestions and criticism during the course of this work.

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Appendix I

Calculation of Torque from Impact Tests Under Rupture

By the use of a series of low speeds, the flywheel of the torsion impact machine can be made to strike the specimen at an energy level that will twist it plastically but not enough to cause failure. In Table V, the results of such a series of tests on 1.10 per cent carbon tool steel, brine quenched 1450 degrees Fahr. (790 degrees Cent.) and drawn at 350 degrees Fahr., are given:

Table V
Impact Tests Below Rupture 1.10 Per Cent Carbon Tool Steel
Brine Quenched 1450 Degrees Fahr., Drawn 350 Degrees Fahr.

Rotor Speed R.P.M.	Rotor Energy In Ft.-Lbs.	Permanent Set	
		Degrees	Radians
120	7	0.5	0.01 (elastic limit)
140	10	2	0.035
200	20	8	0.14
245	30	14	0.25
285	40	20	0.35
350	60	32	0.56
400	80	44	0.77
450	100	56	0.98
700	120 (rupture)	68	1.17

A specimen twisted below its elastic limit will develop a torque directly proportional to the deformation and dependent in magnitude

on the modulus of shear. In Fig. 10, this relation is shown by the line marked "modulus line." If the elastic limit is exceeded, the torque-strain curve departs from this line and plastic flow occurs. Upon release of the load, the specimen will return along a straight line parallel to the modulus line but displaced from it an amount equal to the permanent set. This displacement can be measured and is shown in Table V for each of the rotor speeds employed. The maximum torque and total deformation values reached when the energy of the rotor is completely absorbed must lie somewhere on this elastic recovery line parallel to the modulus. Moreover, the area under the torque-strain curve must equal exactly the original energy of the flywheel.

The first point to be calculated lies on the modulus line itself and is the elastic limit of the steel. It is determined by the lowest energy level of the flywheel that can cause a detectable permanent set in the specimen. The torque and deformation can be evaluated when this energy is known from the following relations.

$$\Phi = \frac{TL}{FJ} \quad \text{EQUATION (2)}$$

Where,

Φ = deformation in radians,

T = torque in lb.-inches,

L = gage length in inches,

F = shear modulus = 10,800,000-lbs. per sq. inch.

J = polar moment of inertia of the section
(.00039 for $\frac{1}{4}$ " diameter).

Evaluating:

$$\Phi = \frac{T}{4100} \quad \text{EQUATION (3)}$$

Referring to Fig. 10, the energy stored in the gage length of the specimen under torsion in terms of torque and deformation can be stated:

$$W = \frac{1}{2} T \Phi = \text{area } OT_{EL}A \quad \text{EQUATION (4)}$$

Where,

W = energy in inch-lbs.

T = torque in lb.-inches,

Φ = deformation in radians.

But,

$$\Phi = \frac{T}{4100} \text{ hence, } W = \frac{1}{2} T \left(\frac{T}{4100} \right) = \frac{T^2}{8200} \quad \text{EQUATION (5)}$$

From the data in Table V:

$$\frac{T_{EL}^2}{8200} = 84 \text{ in. lbs. } T_{EL} = 825 \text{ lb. in. at Elastic Limit.}$$

$$\Phi_{EL} = \frac{T_{EL}}{4100} = 0.20 \text{ radians.}$$

Let, T_{10} = maximum torque developed in the 10-foot-pound test.

The area under the curve between the elastic limit and T_{10} (area $AT_{EL}T_{10}B$, Fig. 10) must be equal to the energy difference $10 - 7 = 3$ foot-pounds (or 36 inch-pounds). If the torque curve be assumed to be a straight line between these points (which introduces a very small error) the product of the average torque between T_{EL} , and T_{10} and the increment of deformation, $\Delta\Phi$ will equal the area under the curve or 36 inch-pounds.

But,

$$\frac{T_{10} + 825}{2} = \text{average torque between elastic limit and } T_{10}.$$

Hence,

$$\frac{T_{10} + 825}{2} \Delta\Phi = 36 \quad \text{EQUATION (6)}$$

If $\Delta\Phi$ can be expressed in terms of T_{10} only, this equation can be solved.

Now,

$$\Delta\Phi = \Phi_{10} - \Phi_{EL}$$

Where,

Φ_{10} = total deformation during 10 ft. lb. test,

and

Φ_{EL} = deformation at the elastic limit = 0.20 radians,

but

Φ_{10} is composed of both elastic and plastic deformation.

The plastic portion is measured after the test, and in Table V is seen to be 0.035 radians. The elastic portion is due to elastic recovery and is dependent on the torque developed, T_{10} according to Equ-

tion (3), $\frac{T_{10}}{4100}$.

Hence,

$$\Phi_{10} = 0.035 + \frac{T_{10}}{4100}$$

Therefore,

$$\Delta\Phi = 0.035 + \frac{T_{10}}{4100} - 0.20 = \frac{T_{10}}{4100} - 0.165.$$

Substituting in Equation (6),

$$\frac{T_{10} + 825}{2} \left(\frac{T_{10}}{4100} - 0.165 \right) = 36.$$

This quadratic is readily solved for T_{10} .

$$T_{10} = 835 \text{ lb. in. } \Phi_{10} = 0.24 \text{ radians.}$$

In a similar fashion,

T_{20} = maximum torque developed in 20 ft.lb. test,

$$\frac{T_{20} + 835}{2} = \text{average torque.}$$

$$\Delta\Phi' = \Phi_{20} - \Phi_{10} = 0.14 + \frac{T_{20}}{4100} - 0.24 = \frac{T_{20}}{4100} - 0.10$$

$$\left(\frac{T_{20} + 835}{2} \right) \left(\frac{T_{20}}{4100} - 0.10 \right) = 120 \text{ in.-lbs.}$$

$$T_{20} = 940 \text{ lb. in. } \Phi_{20} = .37 \text{ radians.}$$

In this way the torque and deformation can be evaluated for each increment, resulting in the complete torque-strain curve shown in Fig. 5.

Appendix II

Derivation of Formula for Stress Calculation

Adapted from Foepl (10)

In a circular section which is loaded in torsion, the total torque developed will be dependent only on the amount of angular twist and the radius of the section. Below the elastic limit, the unit fiber stress at any point is directly proportional to the deformation and also directly proportional to the distance the point lies from the neutral axis. Above the elastic limit, the stress increases much less rapidly with respect to deformation or distance from the axis. As deformation is increased, the elastic limit is first exceeded at the surface fibers and then at points progressively toward the center. Now the total torque set up is the integrated effect of all the stress existing

in the section. Usually torque is plotted against deformation with a constant section radius, but it would be quite possible to study the relation of torque versus radius with a constant angular deformation.

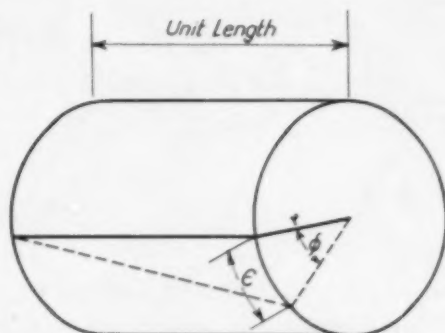


Fig. 11—Relation of Unit Deformation to Angular Deformation.

Referring to Fig. 11, the unit deformation at any point in a section is obtained by Equation (7)

$$\epsilon = \Phi r \quad \text{EQUATION (7)}$$

Where,

- ϵ = unit deformation in inches per inch,
- Φ = angular deformation in radians per inch,
- r = distance of point from neutral axis.

The stress existing at this point is directly dependent on this deformation ϵ , even above the elastic limit. Below the elastic limit, stress is directly proportional to ϵ in a relation expressed by the shearing modulus and above the elastic limit they are no longer proportional, but each are fixed in value when the other is determined.

Now if two sections of different radius, r_1 and r_2 respectively, are deformed so that the stress at the outer fiber of each is the same, and hence ϵ is the same, the following is true,

$$\epsilon_1 = \epsilon_2$$

Therefore,

$$\Phi_1 r_1 = \Phi_2 r_2 \quad \text{EQUATION (8)}$$

The distribution of stress over these two sections will be exactly the same and hence each can be considered to have an average stress,

S_a , acting at the polar radius, $\frac{r}{\sqrt{2}}$. From this, the total torque can be expressed:

$$T = S_a \times \text{section area} \times \text{polar radius} = S_a \cdot \pi r^2 \cdot \frac{r}{\sqrt{2}}$$

$$T_1 = S_a \frac{\pi r_1^3}{\sqrt{2}}$$

$$T_2 = S_a \frac{\pi r_2^3}{\sqrt{2}}$$

$$\text{or } T_1 : T_2 = r_1^3 : r_2^3$$

EQUATION (9)

That is, the ratio of the torques in these two sections possessing the same maximum fiber stress is equal to the ratio of their radii cubed.

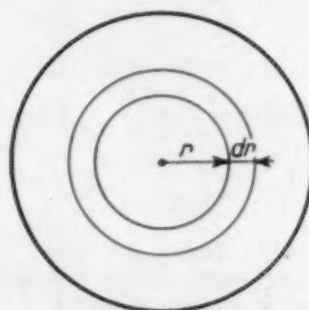


Fig. 12—Relation of Torque and Radius at a Constant Deformation.

If a section is deformed at a constant angle Φ_1 , the relation between the torque and the radius is expressed in a differential form, from Fig. 12:

$$dT = S \cdot 2\pi r^2 dr$$

EQUATION (10)

$$\text{or } \frac{dT}{dr} = 2\pi r^2 S,$$

Where,
 S = unit stress at "r".

Referring again to the sections of r_1 and r_2 , the section of r_1 can be considered to be changing in radius with a constant angular deformation, Φ_1 ; and section r_2 can be considered to be changing in deformation with constant radius, r_2 , such that the maximum fiber stress at the periphery section r_1 is always equal to that at the periphery of r_2 .

Hence from Equations (8) and (9),

$$\Phi_1 r_1 = \Phi_2 r_2,$$

And

$$T_1 : T_2 = r_1^3 : r_2^3$$

of which Φ_1 and r_2 are constants.

Now,

$$\frac{dT_1}{dr_1} = 2\pi r_1^2 S \quad \text{EQUATION (11)}$$

But,

$$T_1 = T_2 \left(\frac{r_1^3}{r_2^3} \right) \text{ and } r_1 = r_2 \left(\frac{\Phi_2}{\Phi_1} \right)$$

Therefore,

$$T_1 = T_2 \left(\frac{\frac{\Phi_2}{r_2} \frac{\Phi_2}{\Phi_1}}{r_2^3} \right)^3 = T_2 \frac{\Phi_2^3}{\Phi_1^3}$$

And,

$$dT_1 = \left(\frac{\Phi_2}{\Phi_1} \right)^3 dT_2 + 3T_2 \frac{\Phi_2^3}{\Phi_1^3} d\Phi_2$$

$$dr_1 = \frac{r_2}{\Phi_1} d\Phi_2$$

Substituting for dT_1 , dr_1 and r_1 in Equation (11)

$$\frac{\left(\frac{\Phi_2}{\Phi_1} \right)^3 dT_2 + 3T_2 \frac{\Phi_2^3}{\Phi_1^3} d\Phi_2}{\frac{r_2^3}{\Phi_1} d\Phi_2} = 2\pi r_2^3 \frac{\Phi_2^3}{\Phi_1^3} S.$$

Simplifying,

$$\Phi_2 \frac{dT_2}{d\Phi_2} + 3T_2 = 2\pi r_2^3 S.$$

$$S = \frac{1}{2\pi r_2^3} \left(\frac{dT_2}{d\Phi_2} \Phi_2 + 3T_2 \right)$$

which can be stated as a general formula for a section of constant radius and varying deformation,

$$S = \frac{1}{2\pi r^3} \left(\frac{dT}{d\Phi} \Phi + 3T. \right)$$

This information can be read from a torque-strain curve and the stress evaluated. In the above equation,

S = Maximum fiber stress in lbs. per sq. in.

r = specimen radius,

T = torque in pound-inches.

Φ = deformation in radians.

$\frac{dT}{d\Phi}$

= slope of the curve at the point selected on the torque-strain curve.

DISCUSSION

Written Discussion: By Robert S. Rose, Vanadium-Alloys Steel Co., Boston.

I propose commenting on two phases of this paper which is a sequel and complementary to those previously reporting results and observations of torsion impact testing.

There has been some doubt expressed in the literature as to the actuality of the high energy values at 375 to 400 degrees Fahr. tempering temperature and lower energy values which obtain at approximately 500 degrees Fahr. tempering temperature. This doubt has its origin in the inability to demonstrate or observe similar variations with tests in tension and flexure. Analysis, however, suggests no reason why identical results should be expected under imposed loads so utterly different. The ratio of shear to normal stress developed determines for a specific material and condition the mode of failure, i.e., the extent and amount of ductility. If this ratio is distinctly different, as it is in torsion, tension, and flexure, no direct correlation should necessarily be obtained. Thus a material in a specific condition may be made to demonstrate variable ductility simply by variation in the type of load. Nevertheless it is natural to look for greater toughness at a lesser hardness and it is so indicated by tests other than in torsion. By increasing the drawing temperature from 400 to 500 degrees Fahr. the hardness drops about 3 R.C. points, the elastic limit is reduced, the shear strength (based upon the definition of hardness) is obviously reduced and yet the ductility (deformation) in torsion is less. Ordinarily a reduction in shear strength is accompanied by an increase in ductility and so it is in torsion except in this critical drawing range. I can conceive of this phenomenon resulting only from an even greater loss of cohesive strength in this drawing range which, in spite of the lesser shear strength, increases the shear to cohesive ratio. An increase in this ratio reduces ductility. Presumably with loads in tension and flexure which begin with much higher ratios the effect is insignificant. Torsion tests consistently indicate loss of toughness upon or immediately following the breakdown of retained solid solutions whether they are ferrite or austenite. For example—stainless iron of low carbon wherein practically no austenite is retained upon quenching shows a pronounced drop in torsion toughness when drawn to temperatures coincident to a change in etching behavior. It appears plausible then to suggest that carbides, either

upon precipitation or soon after, but before reaching visible size, have a deleterious effect upon the cohesive strength of steel.

The authors mention that the torsion impact test does not permit torque-strain measurements. It is because of this limitation that we have preferred the static test which allows evaluation of the components in the toughness product. Messrs. Green and Stout state, however, the torsion impact test more nearly approaches service conditions. Their results, shown in Fig. 1, nevertheless, indicate good agreement between static and impact tests. The shape of curves are identical with a displacement to lower values with static tests. This is precisely the condition Mr. H. Mann at the Watertown Arsenal found in his attempt to correlate energy values of the static and dynamic tension test before consideration was given to energy dissipated in the form of heat. Part of the energy value measured is consumed in deforming the test sample and part is converted to heat. In the static test sufficient time is provided for heat to escape and thus the torque value drops to the level necessary only to deform the sample. The impact test does not allow adequate time and thus the energy values include both deformation and heat requirements. Mann found the heat increment to vary with materials and conditions of the same material and to be at times as high as 25 per cent of total energy value. When this heat value was converted to foot-pounds and added to the static value, direct correlation between static and dynamic tension tests below the transition velocity was obtained. The highest speed of the authors' test is much less than any reported transition velocities in tension. I have not read the authors' paper referred to, reporting their observations of the influence of speed on the torsion impact test, so it is quite possible this factor has already had their consideration. If not, perhaps they will wish to measure its magnitude with the same thoroughness that has been characteristic of their entire series on torsion testing.

Written Discussion: By D. S. Clark, assistant professor of mechanical engineering, California Institute of Technology, Pasadena, Cal.

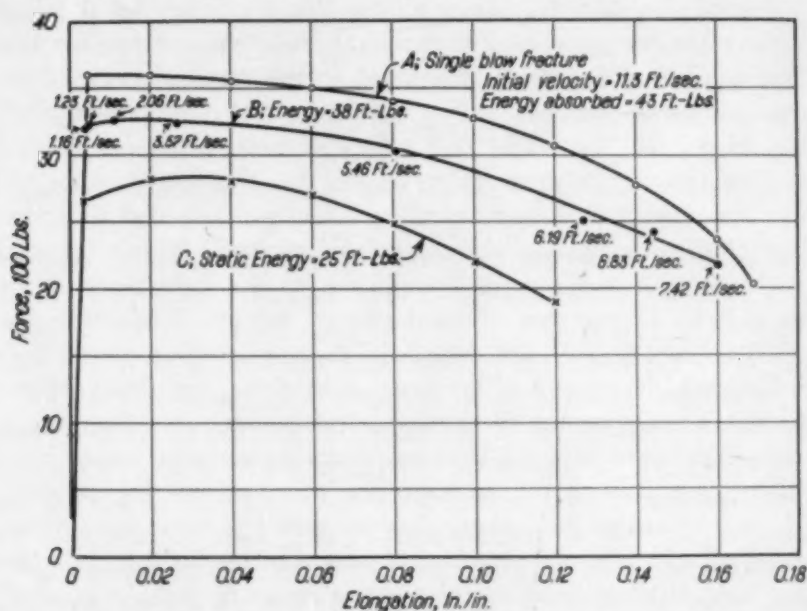
Notched bar testing has for some time maintained an important position on the programs of technical sessions dealing with engineering materials. This condition gives evidence that the study of the behavior of materials which are subjected to rapidly applied loads is of considerable importance. The net result of the work in notched bar testing over a period of several years is an accumulation of a large amount of data which in general cannot be used in a quantitative way. There are so many variables involved in notched bar testing that the true behavior of a material in relation to each variable is difficult to see. Admittedly, it is very useful to know the notch sensitivity of one material relative to another material for the same condition of the notch. The results of the notched bar test are given in terms of the total energy required to bring about the failure of a standard notched bar.

It has been recognized for some time that the energy required to fracture a material is a function of the rate of deformation. Investigations on this subject are referred to as impact testing. Hoyt¹ has very clearly discussed the distinction between notched bar and impact testing. The present paper falls into the category of impact testing, and the authors have chosen the

¹S. L. Hoyt, "Notched-Bar Testing and Impact Testing," *Proceedings, American Society for Testing Materials*, Vol. 38, Part II, 1938, p. 141.

medium of torsion to examine this phenomenon. Others have chosen the tension test. It is encouraging to see that some thought is being given to a study of the stress-deformation relations in torsion impact. It is through these more fundamental types of investigations that progress in understanding the effect of rate of deformation will come about.

When the preprint of the present paper was received it seemed questionable if the stress-deformation relations computed from the data obtained from the subrupture tests were correct. In determining these relations specimens are tested at successively higher rates of deformation in order to subject the



specimens to increasing increments of energy. Therefore each point on the curve is obtained for a different impact velocity. This means that the first point on the curve is obtained at a low velocity, and the last at a velocity about seven times the first. The authors showed in a previous paper² that for the material in question, when tempered at low temperatures, the energy was a function of velocity.

The writer is particularly interested in this phase of the subject because in 1938 he and Dr. G. Dätwyler published a paper³ describing a method which they used to determine stress-strain relations in tension impact. In these tests the forces acting on the specimen are determined with a dynamometer, and the deformations with an appropriate high speed extensometer, in conjunction with a cathode ray oscillograph. In view of this work it does not seem to the writer that the method used by the authors of the present paper will give true stress-deformation relations for velocity-sensitive materials because of the variable velocities involved throughout the tests.

To support this contention the writer, with the assistance of Mr. Alfred Schaff, Jr., conducted a series of impact tests on cold-rolled S.A.E. 1015

²G. V. Luerssen, "The Torsion Impact Properties of Hardened Carbon Tool Steel," *TRANSACTIONS, American Society for Metals*, Vol. 22, 1934, p. 311.

³D. S. Clark and G. Dätwyler, "Stress-Strain Relations Under Tension Impact Loading," *Proceedings, American Society for Testing Materials*, Vol. 38, Part II, 1938, p. 98.

tension specimens with a 0.200-inch diameter and a gage length of 1 inch at different velocities in a pendulum machine. The testing procedure followed in these tests was similar to that used by the authors. Since the mass of the pendulum was constant, different impact energies were obtained by varying the velocity. In these tests the forces and deformations were measured with the equipment developed in the writer's laboratory. The first test consisted of determining the true force at the elastic limit by using a pendulum velocity which would not produce measurable deformation. Specimens were then tested at higher velocities until failure occurred. The results of these tests are given in Curve B of the figure. The velocity for each point is indicated. Another specimen was tested by breaking it with a single blow of 120 foot-pounds. The force-deformation relation for this test is given in Curve A of the figure. Another specimen was tested under true static conditions in which equilibrium conditions were reached for each load applied. The result is shown in Curve C of the figure. These curves indicate that the procedure outlined by the authors when applied to tension impact tests leads to a curve which is above that for static conditions and lower than that for higher velocities. In the test in which an impact energy of 52 foot-pounds (7.42 feet per second) was applied to the specimen, failure occurred with the absorption of 46 foot-pounds of energy. Curve B, according to the method prescribed in the present paper, should give the same energy. However, the area under Curve B corresponds to 38 foot-pounds. This would lead one to believe that Curve B does not represent the true force-deformation relations under all conditions of tension impact.

It must be recognized that the result with the material tested by the authors is not greatly affected by variations of velocity in the range of 100 to 700 revolutions per minute. Under such conditions, the method outlined may give results which are close to true stress-strain relations under torsion impact in this velocity range. This method should be used with the utmost discretion to be sure that the material is not velocity sensitive in the range involved, and that the machine parts are extremely rigid to insure minimum elastic energy in those parts. Correct stress-deformation relations for a velocity-sensitive material might be obtained with this method by using a machine in which the velocity could be made constant and the variation of energy obtained by changing the character of the rotating disk.

The writer would like to inquire if the authors have had any experience with materials other than that reported here, and if so, what the results were.

Fig. 1 in the paper indicates that the material is not as sensitive to velocity changes above 100 revolutions per minute as below 100 revolutions per minute. Has any work been done at velocities above 700 revolutions per minute? One might expect a decrease in energy values after reaching a certain velocity. The writer wonders if the rate of deformation corresponding to 700 revolutions per minute is not near to what might be called the critical velocity for this material.

Written Discussion: A. Bradford Reed, Rolled Thread Die Co., Worcester, Mass.

The privilege of commenting on the latest paper of Mr. Greene and his

associates, interpreting the results of torsion impact testing, is appreciated by the writer who has followed this development since its inception.

I would like to suggest that curves of plastic deformation before rupture and maximum fiber stress be plotted on the same sheet against drawing temperatures. This will enable a tool steel user to select what he considers the best compromise between strength and ductility for the job in hand.

I have never heard "toughness" satisfactorily defined but suggest that it may be expressed in terms of the amount of plastic deformation possible in a given steel at any one of several standard ultimate strengths.

Notwithstanding the above, it should be pointed out that while a very small amount of plastic deformation is desirable in order to help distribute the stress, most tool steel products have failed when this deformation becomes measurable even though the steel has not ruptured. This is because of the close limits now in common use on tools and machine parts.

I would also like to point out that a great many tool steel parts fail, not as the result of a single overload, but after a large number of loadings below the breaking point. I would like to ask if the stress-strain characteristics of a test piece change when loaded repeatedly to a point above the elastic limits but below the breaking point and, also, if a test in which the test piece is ruptured in a single loading is an indication of what can be expected of a test piece subjected to repeated loadings.

Written Discussion: By F. G. Tatnall, Southwark Division, Baldwin-Southwark Corp., Philadelphia.

One of the high points in this paper is the comparison in Fig. 9 of stress-strain curves obtained by static tests and impact tests below rupture, which clearly sets forth information that many people would like to have. This is only one of many interesting facts developed. However, there seems to be an opportunity offered to extend the studies of torsion impact, using the equipment described by Greene and Stout into the controversial field of variable speed impact, which has been previously investigated by Haskell and Mann at Watertown Arsenal and Dr. D. S. Clark at California Institute of Technology. It would seem that the speed of deformation on a very small gage length could be made quite high without altering the equipment or method of testing in torsion impact and that the impact could be applied at various speeds, measuring the absorption of energy at each speed in accordance with methods outlined by Greene and Stout in their subrupture testing until they could find the transition point pointed out by Haskell and Mann, wherein the absorption of energy decreases with increasing striking velocity.

If such means of study could be carried out it would form an interesting means of checking the work that has been done in the past and might throw further light on this subject, which is highly important to manufacturers of projectiles and armor plate. This would be an extremely quick and simple method of determining the effect of heat treatment on the position of Haskell and Mann's transition point.

I would like to ask the authors if they have given consideration to this line of study or whether it is not feasible with the torsion impact method?

The authors referred to the Symposium on Impact Testing in the *Proceedings of American Society for Testing Materials*, Vol. 38, 1938, p. 98. This is

a paper by Dr. Clark and Dätwyler, entitled—"Stress-Strain Relations Under Tension Impact Loading." By an entirely different means these investigators plot stress-strain curves of static tension tests and impact tension, but these show far greater difference between dynamic and static tests, the dynamic elastic limit being double the amount of static in some cases. Valuable points for discussion would be the reasons for these differences of results in different test methods which could form the basis of a complete new investigation, because there is increasing emphasis in testing circles on static versus dynamic testing and the relative value of each.

Oral Discussion

A. V. DE FOREST:⁴ I think it is extremely fortunate that this work on torsion is being carried on in the impact region as well as the other tests which are being carried on in tension. It should be possible to reconcile the two different types of tests and we certainly expect that that is going to come about.

As far as I know, the tension tests have mostly been carried out on material that was in quite a different state from the fully hardened and somewhat drawn high carbon tool steels. There are probably very special relationships in the steels containing residual austenite that do not appear in the steels that primarily have a ferrite base, so that in time the work will have to be extended into both fields. The torsional test has a very great advantage, in that the amount of deformation available is far greater than in the tension test.

One of the difficulties in reconciling any impact test—and perhaps this applies even more to the torsion than to the tensile test—is the temperature at which the tests are carried out. Until we have thermocouples hooked onto cathode ray oscillographs at the point at which rupture is going to take place, we do not know the temperature at which the piece of steel finally comes apart, so that it is quite reasonable to foresee a difference between very rapid loadings and less rapid loadings and the static or semi-static loading that is usually recorded.

The authors are to be congratulated on going after a difficult piece of experimental work and I hope that after they have solved the problem of the smooth test specimen, they will apply the torsional method to a notched specimen.

GEORG SACHS:⁵ I do not understand completely how the authors account for the very great effect of speed on the energy required for breaking the hardened and not tempered specimens. As far as I can see, this may be explained either by an additional formation of martensite during testing at low speed, or by a tempering effect during testing at high speed due to the developed deformation heat which is not conducted away fast enough. I wonder which explanation is favored by the authors.

Authors' Closure

We certainly appreciate the very valuable discussions offered on our paper. Mr. Tatnall's suggestion in regard to the study of critical speeds is worthy

⁴Professor of mechanical engineering, Massachusetts Institute of Technology, Cambridge, Mass.

⁵Assistant professor, department of metallurgy, Case School of Applied Science, Cleveland, O.

of consideration. It appears, however, in order to approach speeds comparable to those used by Mann prohibitive rotor velocities will be required, since the outer fibre of our test section travels in a very limited circle ($\frac{1}{8}$ -inch radius). His comment on the variance of our results with those of Clark and Dätwyler can be explained by noting that the material of their tests was soft, ductile steel, while ours was hardened carbon tool steel; consequently a dissimilarity of behavior might well be expected.

The discussion of Mr. Rose with respect to drawing temperature versus cohesive and shearing strength ratios and resultant ductility is quite plausible. It was shown in a previous paper (Bibliography—Note 4) that drawing about 400 degrees Fahr. entirely transforms the ductile austenitic phase. This is certain to have a marked influence on the cohesive and shearing strength characteristics of the steel.

Mr. Reed's suggestion to plot permanent deformation and strength data on one chart is very good. However, in the paper it was found more convenient to present the data as shown, using identical co-ordinates for all of the drawing temperatures.

Repeated loading above the elastic limit alters the stress-strain characteristics, at least for drawing temperatures below 300 degrees Fahr. If a specimen is subjected to a subrupture test by using a rotor speed less than necessary for breaking, and then retested in this fashion until failure occurs, the total energy absorption and ductility will be much reduced from that of a single test. This absorption, together with the low ductility displayed by specimens in the 15-minute static tests, may be explained by the austenitic breakdown as discussed in paper (Bibliography—Note 9). Mr. Reed's last question enters the realm of fatigue testing, and it is believed that the torsion impact test has as much value as any other single-load physical test.

Dr. Clark's remarks are very scholarly and it is difficult to answer his discussion without reference to the earlier paper. (Bibliography—Note 2). In this paper the emphasis was placed upon mechanical conditions rather than speed. At that time it was felt that the rotor should be given such an initial speed as to have at least a 25 per cent residual energy following the test so that the rupture would occur under definitely impact conditions. A more thorough study of speed was reported in Bibliographical Note 9, and has been summarized in Fig. 1 of the present paper. From these results it was considered justifiable to use subrupture tests for stress-strain analyses without fear of unknown velocity effects.

No other materials have been subjected to this type of analysis with the exception of high speed steel which shows a toughness peak at a draw of 900 degrees Fahr. and a rapid decrease of torsion impact values as the secondary hardening range is reached.

With respect to high velocity testing, check results have been obtained up to rotor speeds of 2500 revolutions per minute. It is evident, of course, that at higher speeds, the increments on the tachometer represent greater energy values, and hence the possibility of error in reading becomes correspondingly more serious. This could be remedied by refitting the torsion impact machine with a lighter rotor. However, as stated above, the linear velocities of the torsion impact are much below those employed by Mann.

The high energy absorption of soft ductile materials in impact as compared to tests under static conditions is well established. There is no evidence of which the authors are aware that demonstrates that this relation is true to the same degree in hardened carbon tool steel. It will be noticed in Fig. 9 of our paper that the impact stress and strain values are consistently above those of static tests; hence there is no indication that a variance of results necessarily exists between these two studies.

The subject of testing the effect of notches mentioned by Professor de Forest is now under investigation. The matter of temperature at the time of rupture deserves serious attention. In torsion impact testing the ruptured section of the specimen usually becomes uncomfortably warm.

Professor Sach's question can be answered by assuming the breakdown of retained austenite is caused by the cold work. This has been discussed fully in a previous paper (Bibliography—Note 9) and is to be the subject of a future study.

THE SPHEROIDIZING OF STEEL BY ISOTHERMAL TRANSFORMATION

BY P. PAYSON, W. L. HODAPP AND J. LEEDER

Abstract

Although many recent discussions on the nomenclature of the aggregates of ferrite and carbide have stressed the formation of lamellar pearlite as the primary product of the transformation of austenite at high temperatures, the present authors show that spheroidal carbides are formed directly from austenite, if the temperatures to which the austenite is heated, and at which it is subsequently allowed to transform, are carefully controlled. The steels studied are plain carbon steels with carbon varying from 0.33 to 1.16 per cent.

New procedures for spheroidizing steels have been based on these observations, and practical annealing cycles requiring much shorter furnace times than those now in use have been developed. The new cycles involve heating to a temperature over the critical for the steel, cooling at a convenient rate to a temperature not far below this critical, holding for a necessary time at this temperature, then finish cooling in air. The total furnace times required, for small furnace charges, are about 6 to 20 hours, depending on the grade of steel.

DURING an investigation of the annealing of a sample of S.A.E. 4345 steel, it was observed that when this steel was cooled from 1450 to 1275 degrees Fahr. (790 to 690 degrees Cent.), and held at the latter temperature for an appreciable time, some transformation took place, and the product was ferrite and spheroids of carbide, as shown in Fig. 1. This was quite surprising because several papers (1)¹ recently had emphasized that austenite transforms only to lamellar pearlite at temperatures not far below the critical.²

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

²Throughout the discussion of the work done at the authors' laboratory, the "critical" will be used to refer only to the temperature at which austenite begins to form when the steel is heated. The authors believe that the expression "critical temperature on cooling" is definite only in connection with a specified cooling rate.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, P. Payson is in charge of the research laboratory, W. L. Hodapp is associated with the research laboratory, Crucible Steel Company of America, Harrison, N. J., and J. Leeder is a graduate student at Massachusetts Institute of Technology, Cambridge, Mass. Manuscript received June 26, 1939.



Fig. 1—S.A.E. 4345 Steel, Heated to 1450 Degrees Fahr., Transferred to Lead Bath at 1275 Degrees Fahr., Held at This Temperature for 48 Hours, and Quenched in Water. Structure Shows Spheroidite Formed Direct From Austenite, and Martensite (Smooth White Areas). Etched in Nital-Picral. $\times 1000$.

It is true that Grossmann (2) had shown that S.A.E.-52100 may form "spheroidized pearlite" during the very slow cooling of this steel from 1400 degrees Fahr. (760 degrees Cent.), but he implied that the composition of the steel together with details of the cooling cycle were responsible for this exception to the rule. Since the production of a spheroidal structure in tool steels is of considerable practical importance, and since the established practice (3) involves annealing cycles which are time consuming, it seemed worthwhile to study the conditions under which austenite transformed to a spheroidal product. Aside from its academic interest, such an investigation held forth a promise of the development of an annealing procedure which might produce the desirable spheroidal product in a shorter time than was necessary in the customary spheroidizing anneal.

EXPERIMENTAL PROCEDURE

Although all investigations of the behavior of steel during annealing have involved the use of various rates of cooling, it occurred to the authors that annealing problems could be studied more

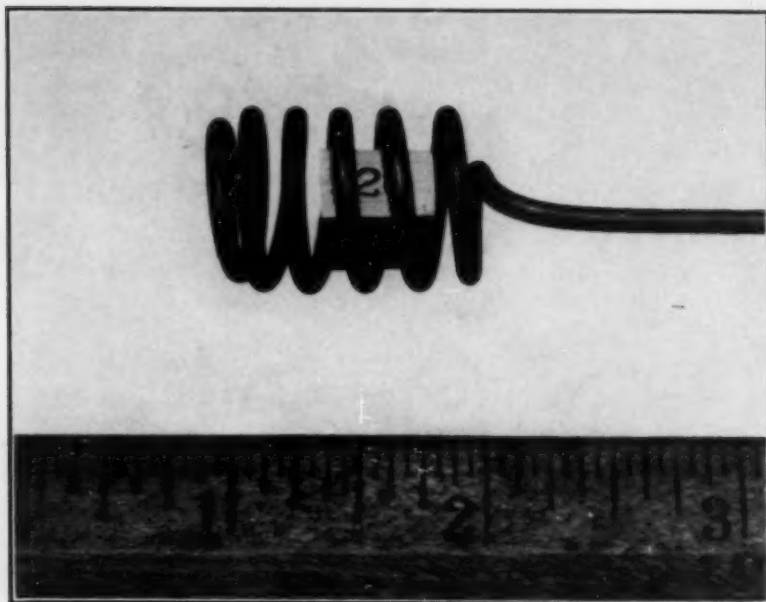


Fig. 2—"Spiral" for Handling Small Test Pieces.

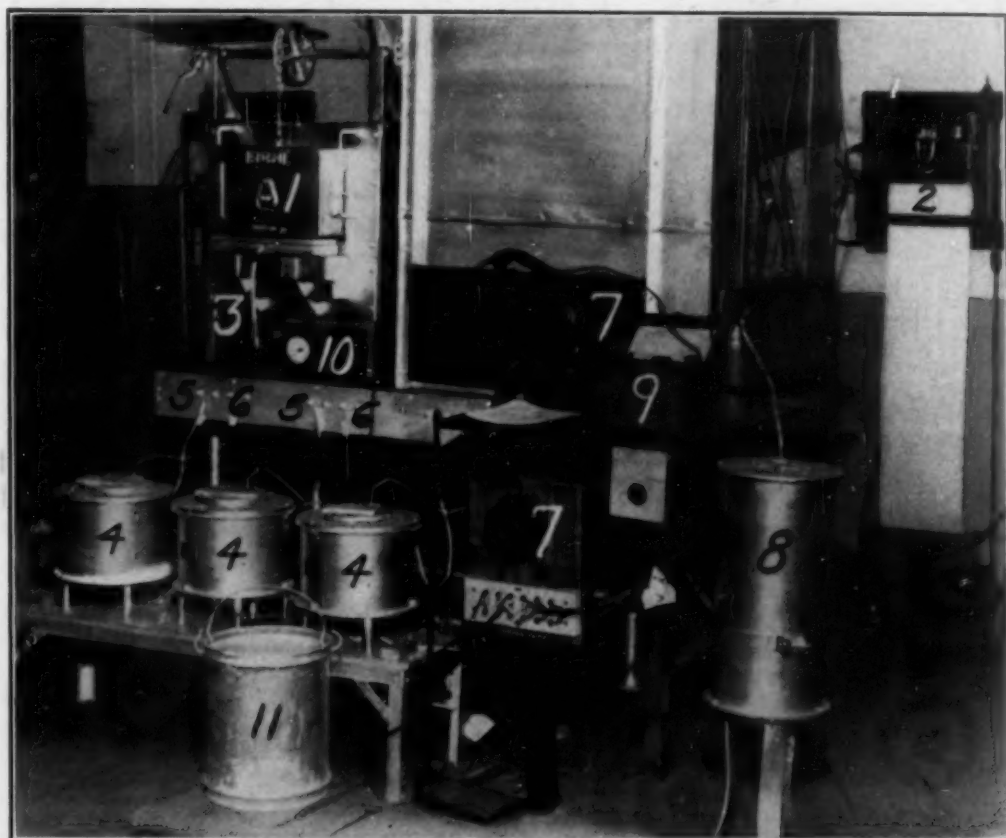


Fig. 3—Complete Set-up for Isothermal Transformation Work.

- | | |
|--|------------------------------------|
| 1. Heating Furnace. | 6. Check Couples for Pot Furnaces. |
| 2. Controller for Heating Furnace. | 7. Multiple Switches for Couples. |
| 3. Check Couple in Const. Temp. Box in Furnace. | 8. Furnace for Checking Couples. |
| 4. Pot Furnaces for Lead, Salt and Low M.P. Alloy Baths. | 9. Semi Precision Potentiometer. |
| 5. Control Couples for Pot Furnaces. | 10. Timer. |
| | 11. Quenching Tank. |

logically by means of isothermal transformations. For this type of work, small pieces about $\frac{1}{2} \times \frac{3}{8} \times \frac{3}{8}$ inch, or smaller, are used. These are heated in an electric furnace with very close temperature control for "austenitizing"; they are then "transformed" either in the same furnace, or in another furnace, or in lead, or salt baths, for varying periods of time; and are finally quenched in water. The quenched samples are examined for structure so that the amount of transformation and the type of transformation product can be determined for the particular set of conditions.

To facilitate the handling of these small pieces, a spiral loop at the end of, and integral with, a 0.095-inch wire of Rezistal 7 is used, as illustrated in Fig. 2. The pieces are readily put into, and removed from, the "spirals"; the wire is of sufficient stiffness to allow the rapid transfer of the pieces from furnace to bath; the spirals offer practically no hindrance to the thorough quenching of the piece; and the wires, being scale resistant and austenitic, can be used for a very large number of heatings and quenchings, without serious deterioration.

A "constant temperature box" is used in the electric muffle furnace to provide a space practically free from temperature fluctuations between the "on" and "off" periods in the automatic control operation. This is merely a small rectangular welded Rezistal 7 box. A check thermocouple placed in the box and connected to a semiprecision potentiometer gives an accurate measure of the temperature in the box. The general setup for this work is shown in Fig. 3. Temperatures are controlled and measured to about plus or minus 3 degrees Fahr. (1.7 degrees Cent.). The equipment is used to establish critical temperatures as well as to develop data for transformation temperature-time curves.

MATERIALS INVESTIGATED

The materials used throughout this investigation were of regular commercial manufacture. A large number of heats, more than twenty-five, were used to check some of the findings, but most of the experimental work was done on $\frac{3}{4} \times \frac{3}{8}$ -inch forged bars from only a few of these. See Table I for steel composition.

ELIMINATION OF PEARLITE FROM ORIGINAL STRUCTURES

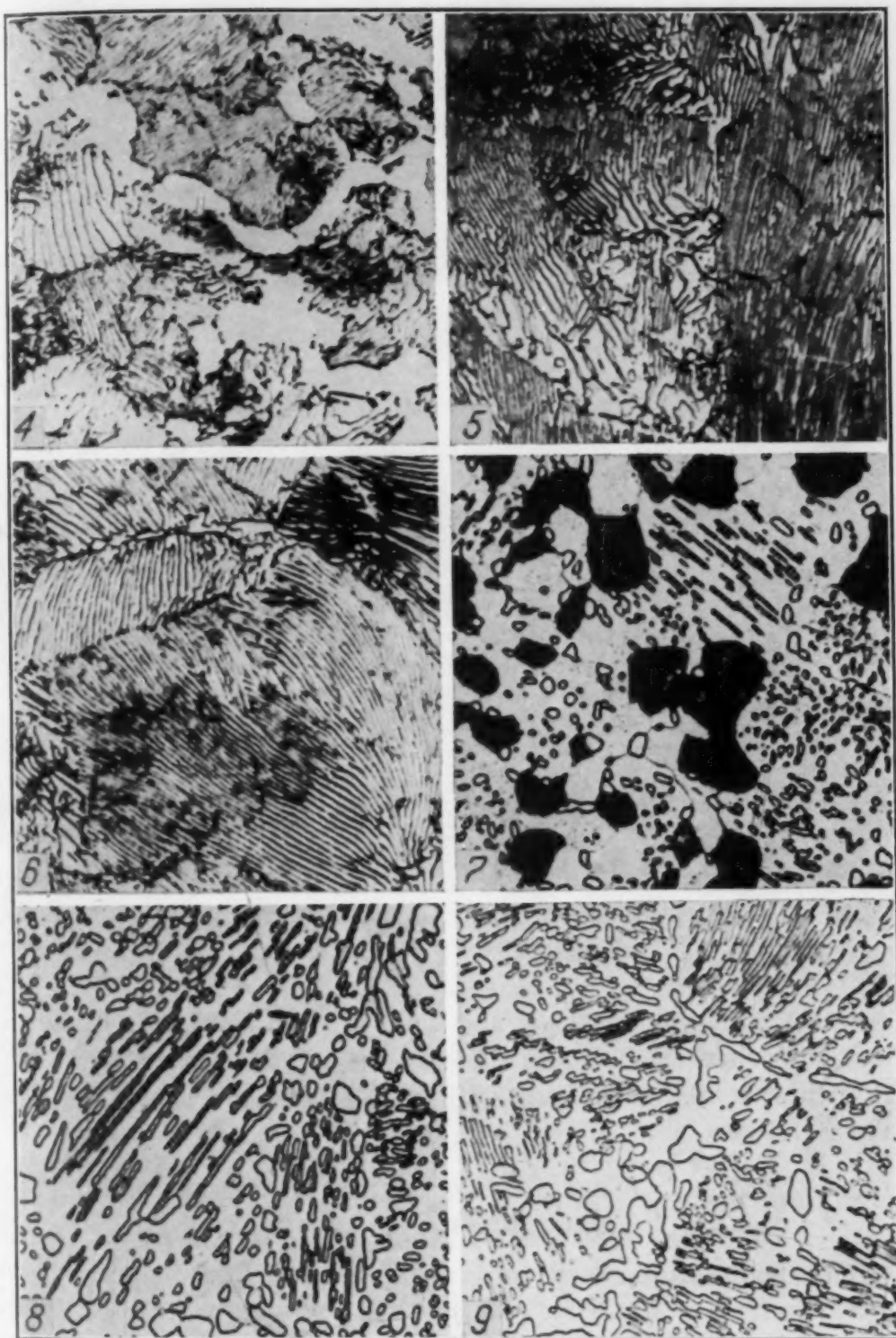
The "as forged" structures—hereinafter referred to as "original" structures—of all the bars showed fine to medium pearlite,

Table I
Analyses of Steels Studied Thoroughly

No.	Per Cent					Melting
	C	Mn	Si	Ni	Cr	
A	0.33	0.34	0.70	0.12	0.17	Induction
B	0.46	0.80	0.19	0.17	0.13	Open-Hearth
C	0.63	0.18	0.25	0.08	0.08	Arc
D	0.85	0.23	0.18	0.05	0.07	Arc
E	1.16	0.27	0.21	0.08	0.11	Arc

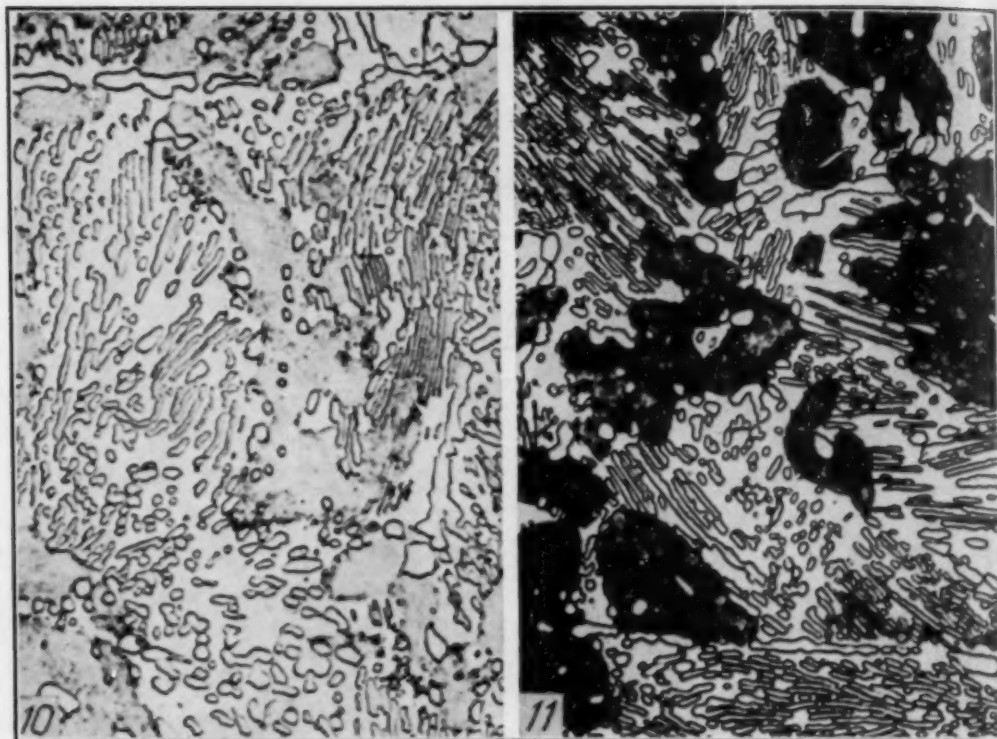
with proeutectoid ferrite in the lower carbon steels, and proeutectoid cementite in the 1.16 per cent carbon steel. Some of these are illustrated in Figs. 4, 5 and 6. To establish the effect on the pearlite of prolonged heating at a temperature below the critical, all steels were heated at 1340 degrees Fahr. (725 degrees Cent.) for 50 hours and water-quenched. After this "spheroidizing" treatment, all steels showed some residual lamellar structure as illustrated in Figs. 7, 8 and 9. (Steel B, because of its higher manganese, showed an appreciable amount of austenite formed after this treatment.) Another set of samples was heated at 1370 degrees Fahr. (745 degrees Cent.) for 16 hours, and water-quenched. These showed martensite in every case, establishing that this heating was over the critical, but although most of the steels were now without a trace of lamellar carbides, the hypereutectoid steel still showed an appreciable amount of pearlite, as in Fig. 10. With an ordinary etch it may be difficult to distinguish the ferrite from the martensite in such a structure. However, as is well known, a short tempering at about 750 to 1000 degrees Fahr. (400 to 540 degrees Cent.) will cause the areas which were martensite to etch dark, and thus be readily differentiated from the ferrite. This procedure was followed on the sample of Fig. 10 and the structure shown in Fig. 11 was obtained. Throughout this investigation a tempering procedure was used to permit easy recognition of the amount of austenite present in the steel when the sample was quenched, and it will be understood, therefore, that the dark etching areas in all the photomicrographs represent austenite that has been quenched and tempered.

In order to determine whether or not the transformation product contained lamellar pearlite, it was essential first to heat the steels to temperatures at which all the lamellar carbides from the original structure would be eliminated. All steels were therefore examined after 1-hour heatings at increasing 10 degrees Fahr. intervals from 1360 degrees Fahr. (740 degrees Cent.), to establish "minimum



Figs. 4, 5 and 6—Original Structure Steels B, D and E with 0.46, 0.85 and 1.16 Per Cent Carbon Respectively.

Figs. 7, 8 and 9—Structures After Long Time Subcritical Spheroidizing Treatment of the Above Steels. 50 Hours at 1340 Degrees Fahr. All Had Picral Etch. $\times 1000$.



Figs. 10 and 11—Residual Lamellar Carbides in Steel E—1.16 Per Cent Carbon, After 16 Hours at 1370 Degrees Fahr.

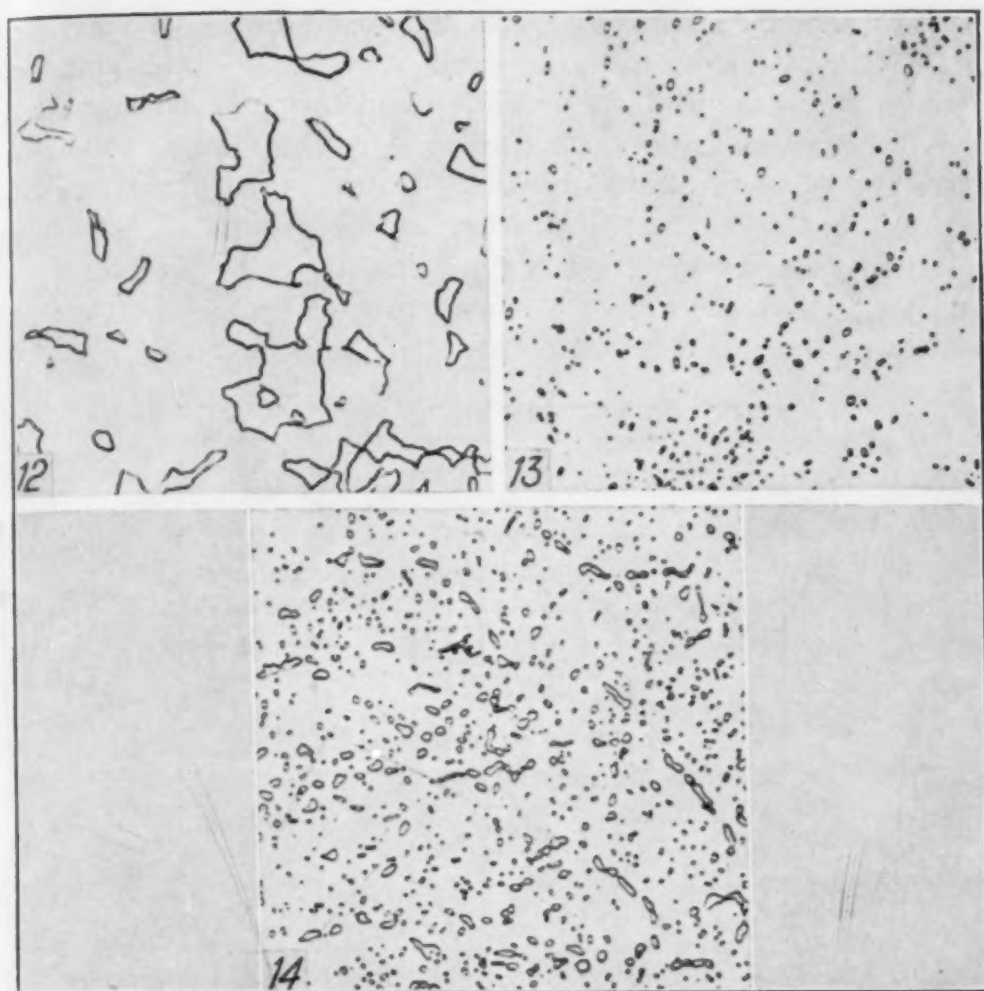
Fig. 10—As Quenched—Martensite Plus Residual Lamellar Carbides in Ferrite.

Fig. 11—Same as Fig. 10 But Former Martensitic Areas Darkened by Tempering.

austenitizing" temperatures, or the lowest temperatures at which the steels would consist of austenite plus nonlamellar carbides only (with or without proeutectoid ferrite, depending on the chemical analysis). These minimum austenitizing temperatures were found to be 1370 degrees Fahr. (745 degrees Cent.) for Steel B; 1380 degrees Fahr. (750 degrees Cent.) for Steels C and D; 1390 degrees Fahr. (755 degrees Cent.) for Steel A; and 1400 degrees Fahr. (760 degrees Cent.) for Steel E. The structures of three of the steels as quenched from these minimum austenitizing temperatures are shown in Figs. 12, 13 and 14.

FACTORS AFFECTING PRODUCTION OF SPHEROIDAL TRANSFORMATION PRODUCT

In their earlier work on S.A.E.-4345 steel, the authors had found that when this steel was allowed to transform at 1275 degrees Fahr. (690 degrees Cent.), the product of transformation was spheroidal, whereas when the steel transformed at 1250 degrees Fahr. (675 degrees Cent.) the product contained lamellar pearlite. It was to be expected then that if a spheroidal product was to be



Figs. 12, 13, 14—Structures of Steels B, D and E with 0.46, 0.85 and 1.16 Per Cent Respectively Quenched After 1 Hour at "Minimum Austenitizing" Temperatures. Specimens Quenched From 1370, 1380 and 1400 Degrees Fahr. Respectively. Picral Etch. $\times 1000$.

obtained at all, it would be found at high transformation temperatures. Furthermore, experience had shown that in the usual annealing operations, spheroidal products were obtained only when very slow cooling rates were used. The slower the cooling rates, the longer the sojourn at each temperature during cooling, and the inference from this was that fairly long times must be allowed for the transformation to take place at high temperatures. To establish the times required for the completion of transformation at various temperatures, transformation temperature-time curves for the upper part of the transformation region were obtained for these steels by the method taught by Davenport and Bain (1a). Samples were heated to the minimum austenitizing temperatures, then transferred

to lead baths maintained at constant temperatures, held here for various time periods, then quenched in water. These quenched samples were examined for structure, and they showed quite definitely whether transformation had occurred, how much had taken place, and what the transformation product was. From the data thus obtained curves were plotted, after the manner of Davenport and Bain, to show on a logarithmic time scale the beginning and end of transformation at each temperature. The high temperature portions of some of these curves are shown in Fig. 15.

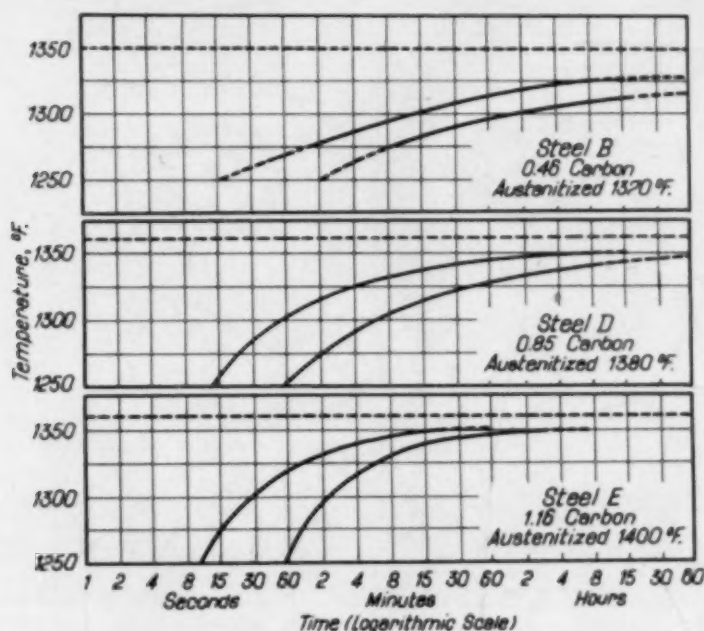
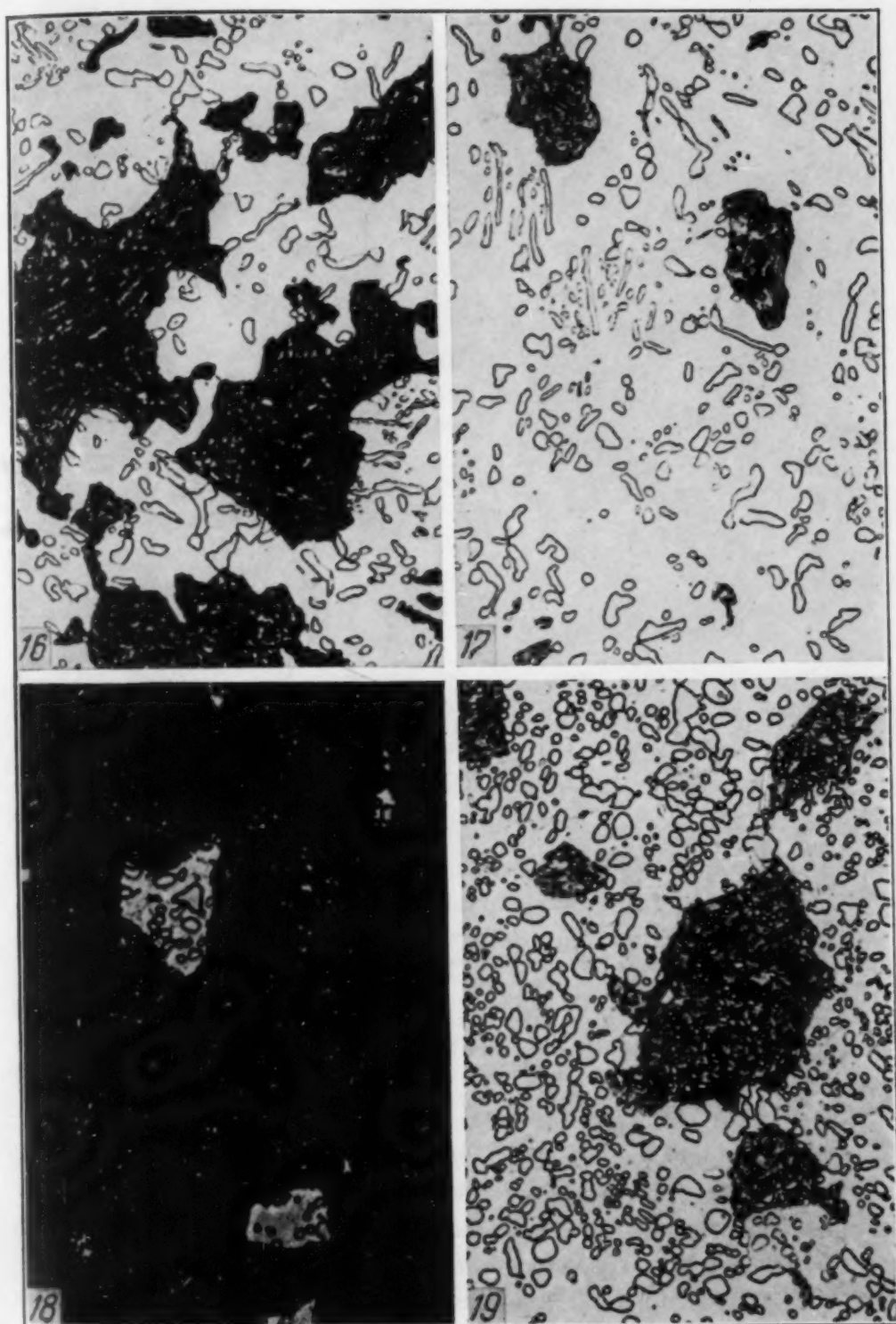
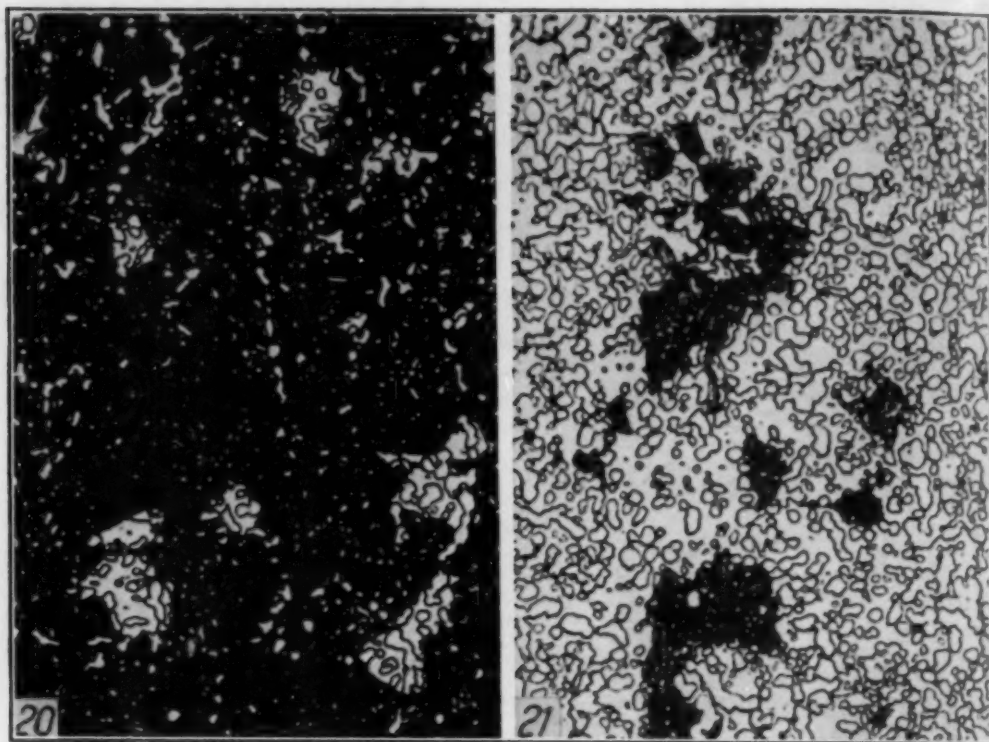


Fig. 15—Upper Portions of Transformation Temperature-Time Curves of Steels B, D and E for Minimum Austenitizing Temperatures.

It was found, as shown in Fig. 15, that the rates of transformation at these high temperatures were much slower in some steels than in others, but all steels transformed directly to spheroidal carbides at the highest transformation temperatures; to lamellar pearlite at the lower temperatures, and to mixtures of spheroidal and lamellar pearlite at intermediate temperatures. Contrary to the classic opinion that lamellar pearlite forms first from austenite, and then spheroidizes, or divorces, during the subsequent slow cooling, it was found that the first areas to transform at the higher temperature showed spheroids without any indication that these spheroids had developed from a prior lamellar structure. Figs. 16 to 21 show the early and late stages in the transformation of some of these steels



Figs. 16, 17, 18 and 19—Direct Formation of Spheroidite From Austenite in Steel B (0.46 Per Cent Carbon) and Steel D (0.85 Per Cent Carbon). Austenitized at 1370 and 1380 Degrees Fahr. Respectively for 1 Hour. Figs. 16 and 17 Transformed at 1310 Degrees Fahr. for 4 and 8 Hours Respectively. Figs. 18 and 19 Transformed at 1335 Degrees Fahr. for 2.5 and 4 Hours Respectively. All Specimens Water-Quenched and Tempered. Figs. 16 and 17 Picral Etch. Figs. 18 and 19 Nital Etch. $\times 1000$.



Figs. 20 and 21—Direct Formation of Spheroidite From Austenite in Steel E, 1.16 Per Cent Carbon Austenitized at 1400 Degrees Fahr. for 1 Hour. Fig. 20—Transformed at 1330 Degrees Fahr. for 3 Minutes. Fig. 21—Transformed at Same Temperature for 10 Minutes. Specimens Water-Quenched and Tempered. Nital Etch. $\times 1000$.

at the higher temperatures. These observations lead to the conclusion that, just as lamellar pearlite is the primary transformation product of austenite under certain conditions, and bainite is the product of transformation under different conditions, and martensite under still other conditions, so spheroidite may be the direct product of the transformation of austenite under a specific set of conditions. These conditions will be recognized as the discussion proceeds.

The range of transformation temperatures in which only a spheroidal product would form was quite limited for some steels, and fairly wide for others. Table II shows for the five steels the amounts of lamellar pearlite in the transformation products formed at various temperatures after the steels had first been heated to their respective minimum austenitizing temperatures. The two higher carbon steels, D and E, are relatively easy to spheroidize because their transformation rates are rather fast and the product is spheroidal over a fairly wide range of transformation temperatures.

Table II
Approximate Percentage of Lamellar Pearlite in Product Formed at Different Transformation Temperatures When Steels Are First Heated to Minimum Austenitizing Temperatures

Steel Per Cent Carbon	Austenitizing Temp. °F. °C.	Transformation Temperatures						
		1355°F. 735°C.	1350°F. 732°C.	1340°F. 727°C.	1330°F. 721°C.	1320°F. 716°C.	1310°F. 710°C.	1300°F. 704°C.
Percentage of Lamellar Pearlite								
A—0.33	1390	754	15%	75%	90%
B—0.46	1370	743	10%
C—0.63	1380	749	40%	90%	100%
D—0.85	1380	749	tr.	15%	30%	40%
E—1.16	1400	760	tr.	5%

The lower carbon steels are rather slow to transform at the temperatures where only a spheroidal product is formed, and the range of temperatures in which the transformation product is spheroidal is quite narrow; these steels are therefore more difficult to spheroidize completely. It was found in these experiments that the formation of a spheroidal product could be made easier by a short heating—two hours or so—at a temperature just below the critical, say 1320 to 1340 degrees Fahr. (715 to 725 degrees Cent.), before the steel was austenitized. The subcritical preheating treatment increased the transformation rate of the austenite, and decreased the amount of lamellar, or elongated carbides, in the transformation product. This treatment is especially beneficial for the lower carbon steels which have rather slow transformation rates, and which have a greater tendency, than the higher carbon steels, to

Table III
Annealing Procedures for Producing Nonlamellar Annealed Structures in Carbon Steels

Steel	Procedure	Amount of Lam. Pearlite	Hardness	
			Brinell	Rock. B
A—0.33 Per Cent Carbon	Heat 2 hrs. at 1340° F. (727° C.), raise temp. to 1390° F. (754° C.), hold 1 hr.; cool to 1350° F. (732° C.), hold 8 hours, cool to 1340° F. (727° C.), hold 4 hours; finish cool in air.	less than 10%	126	76
B—0.46 Per Cent Carbon	Heat 2 hrs. at 1340° F. (727° C.), raise temp. to 1370° F. (743° C.), hold 1 hr.; cool to 1310° F. (710° C.), hold 8 hours; finish cool in air.	less than 1%	134	79
C—0.63 Per Cent Carbon	Heat 2 hrs. at 1340° F. (727° C.), raise temp. to 1380° F. (749° C.), hold 1 hr.; cool to 1340° F. (727° C.); hold 8 hours; cool to 1330° F. (721° C.), hold 2 hours; finish cool in air.	less than 5%	128	76
D—0.85 Per Cent Carbon	Heat to 1380° F. (749° C.), hold 1 hr.; cool to 1335° F. (724° C.), hold 6 hrs.; finish cool in air.	trace	143	81
E—1.16 Per Cent Carbon	Heat to 1400° F. (760° C.), hold 1 hr.; cool to 1335° F. (724° C.), hold 3 hrs.; finish cool in air.	none	174	90

form lamellar, or elongated carbides. By use of this preheat, it is possible to get, in plain carbon steels, annealed structures free from lamellar pearlite, or nearly so, in reasonably short times—about 6 to 20 hours. Details of the annealing procedure, together with the resulting hardnesses and percentages of pearlite in the structure, are given in Table III.

In order for these procedures to give satisfactory results, especially for steels like A, B and C, it is necessary that all temperatures be very closely regulated, and where it is possible, it is desirable that the temperatures be measured by thermocouples in contact with the steel.

For commercial annealing, it is desirable to have as much leeway as possible for both the austenitizing temperature and the transformation temperature. To establish the maximum temperature deviations permissible, samples of all steels were heated to temperatures higher than their minimum austenitizing temperatures, and then transformed at the optimum transformation temperature for the formation of a spheroidal product, as shown in Table II. It was found that for Steels A, B and C the amount of lamellar pearlite in the transformation product increased at once as the austenitizing temperature was raised, and the transformation rates at the high temperatures were also appreciably decreased. Steel D showed only a small amount of lamellar pearlite when it was transformed at 1340 degrees Fahr. (725 degrees Cent.) after a heating as high as 1450 degrees Fahr. (790 degrees Cent.), and about 50 per cent lamellar pearlite after a heating at 1470 degrees Fahr. (800 degrees Cent.). The hypereutectoid steel, E, could be heated as high as 1580 degrees Fahr. (860 degrees Cent.) without forming lamellar pearlite when it was transformed at 1340 degrees Fahr. (725 degrees Cent.), and could be heated as high as 1450 degrees Fahr. (790 degrees Cent.) without forming much lamellar pearlite when it was transformed at 1320 degrees Fahr. (715 degrees Cent.).

These data, showing the appreciable variations in both the austenitizing and transformation temperatures within which it is possible to produce a completely spheroidal structure in the hypereutectoid steel, make it clear why the industry has always found this type of steel easy to spheroidize, and the lower carbon steels more difficult.

Another observation of practical significance was made, and that is, that prolonged heating at a lower austenitizing temperature is equivalent to a shorter heating at a higher austenitizing temperature,

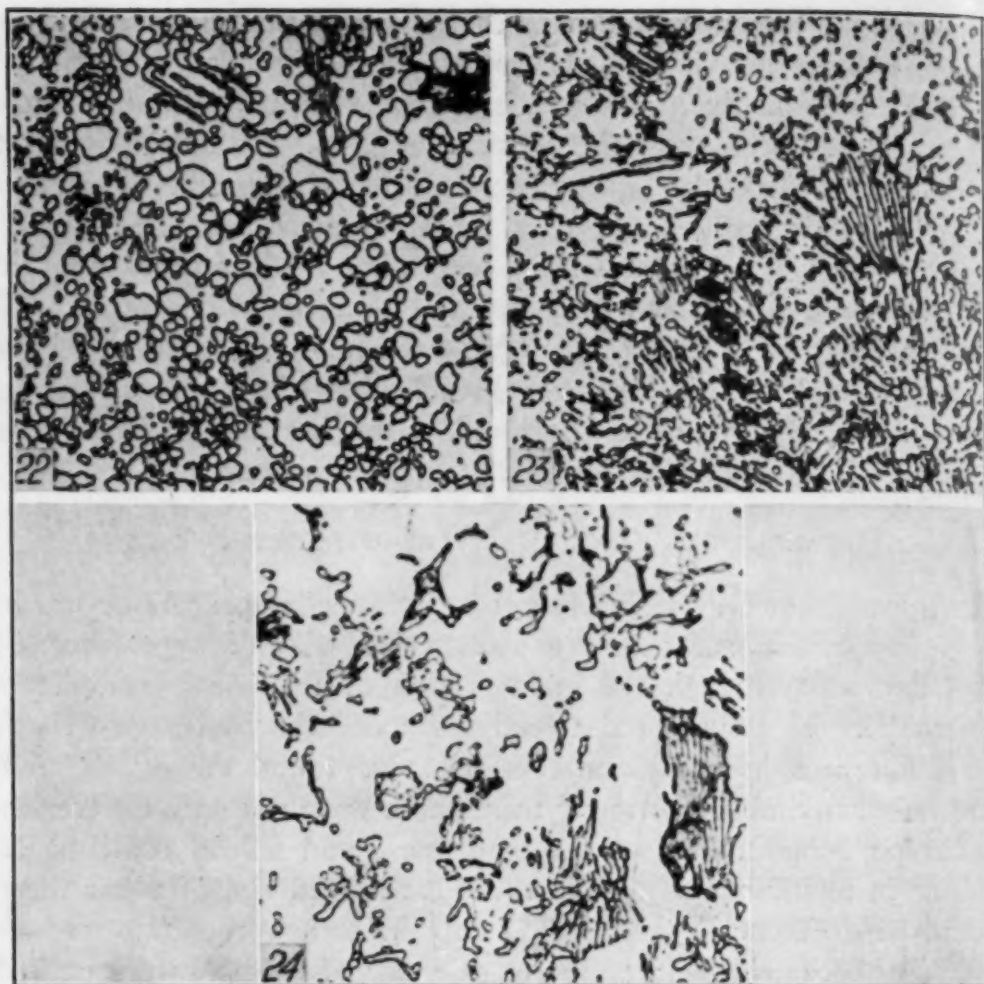
in regards to the subsequent rate of transformation, and the formation of lamellar pearlite. This is especially true of steels like A, B, and C, which are sensitive to changes in austenitizing temperatures, but it is also found in the much less sensitive Steel E. Although the latter is not much affected by heating times up to 16 hours at 1400 degrees Fahr. (760 degrees Cent.), at a heating temperature of 1450 degrees Fahr. (790 degrees Cent.) increasing holding times affected the formation of pearlite when the steel was transformed at 1320 degrees Fahr. (715 degrees Cent.), to the following degree: 1 hour heating time, less than 1 per cent lamellar pearlite; 4 hour heating, about 5 per cent lamellar pearlite, 8 hour heating, about 15 per cent; and 16 hour heating, about 25 per cent lamellar pearlite.

DECARBURIZED ZONE IN HYPEREUTECTOID STEELS

It has frequently been observed that lamellar pearlite occurs in the lower carbon zones in the decarburized rim of hypereutectoid steel bars even though the main section of the bar is completely spheroidal. This suggested a study of a decarburized bar to check the differences in transformation behavior found among the five steels discussed above. Among these there were not only differences in carbon content, but also in manganese and silicon contents, as well as in melting practices; in the decarburized rim of a bar there would be differences, presumably, only in carbon content.

For this purpose, some pieces of a bar of Steel E were packed in wet clay in a pipe and heated at 1500 degrees Fahr. (815 degrees Cent.) for about 16 hours. This treatment resulted in a fairly uniform decarburized layer about 0.030 inch deep.

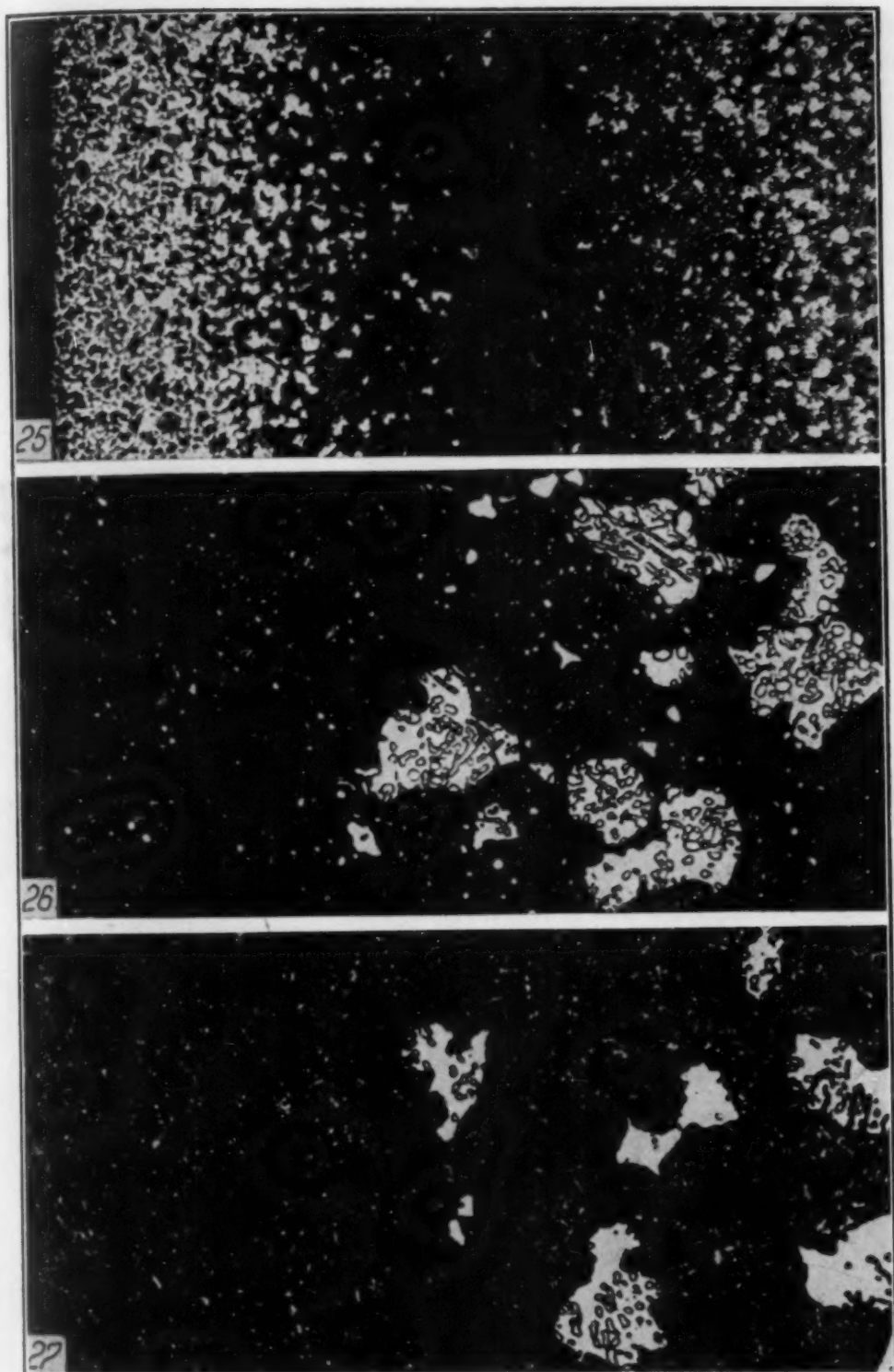
In the first experiment with the decarburized steel, a piece was heated at 1410 degrees Fahr. (765 degrees Cent.), and then transformed at 1300 degrees Fahr. (705 degrees Cent.). This showed an almost completely spheroidal product in the core, and principally lamellar pearlite in the lower carbon regions of the rim, as shown in Figs. 22, 23 and 24. Thus it was established that, with a given set of conditions, the lower carbon material transformed to lamellar pearlite, even though the higher carbon material transformed to spheroidite. In another experiment, a decarburized piece was heated to 1410 degrees Fahr. (765 degrees Cent.), transferred to a lead bath at 1325 degrees Fahr. (720 degrees Cent.), held for 2.5 minutes, and then quenched in water and tempered. This section showed an appreciable amount of transformation both in the



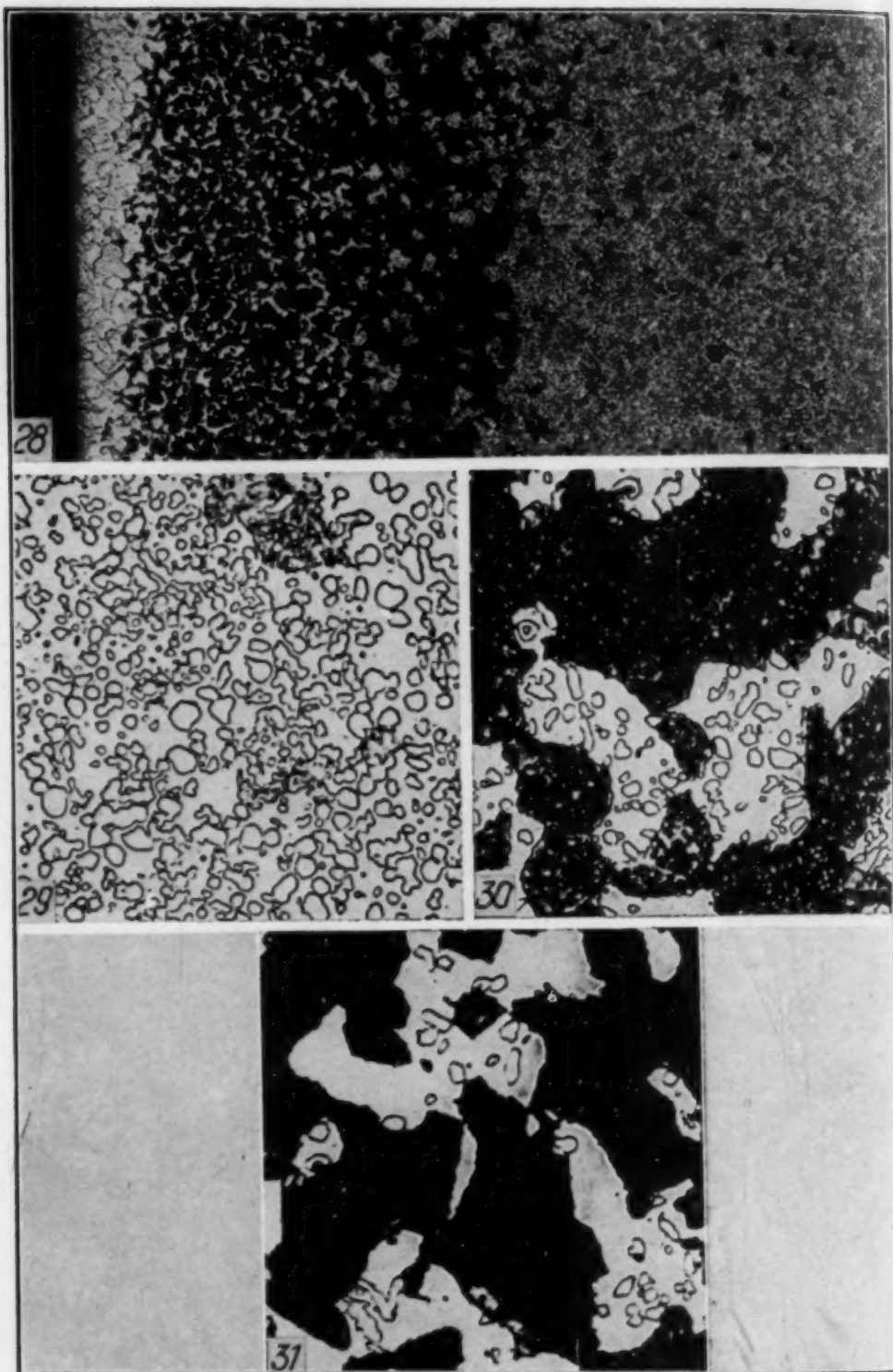
Figs. 22, 23, 24—Structures in Decarburized Bar of Steel E, 1.16 Per Cent Carbon Austenitized at 1410 Degrees Fahr., and Transformed Completely at 1300 Degrees Fahr. Fig. 22—Core of Bar. Fig. 23—Medium Carbon Zone in Decarburized Rim. Fig. 24—Low Carbon Zone in Decarburized Rim 0.007 Inch From Edge. Picral Etch. $\times 1000$.

highest carbon and lowest carbon zones but practically no transformation in the region of intermediate carbon content, as shown in Figs. 25, 26 and 27. Steels of intermediate carbon content are thus shown to have slower transformation rates than either the low carbon hypoeutectoid, or high carbon hypereutectoid, steels.

Finally, a decarburized sample was heated at 1410 degrees Fahr. (765 degrees Cent.), transferred to a lead bath at 1350 degrees Fahr. (730 degrees Cent.), held here for 3 hours, then quenched in water, and tempered. This section showed that under these conditions, all parts of the bar transformed to a spheroidal product. However, the rates of transformation in the lower carbon regions were much slower than that in the original high carbon



Figs. 25, 26, 27—Structures in Decarburized Bar of Steel E, 1.16 Per Cent Carbon Austenitized at 1410 Degrees Fahr., Transformed at 1325 Degrees Fahr. for 2.5 Minutes, Water-Quenched and Tempered. Fig. 25—Edge of Sample. $\times 100$. Fig. 26—Core to Medium Carbon Zone 0.023 Inch From Edge. $\times 1000$. Fig. 27—Low Carbon Zone 0.009 Inch From Edge. Nital Etch. $\times 1000$.



Figs. 28, 29, 30, 31—Structures in Decarburized Bar of Steel E, 1.16 Per Cent Carbon Austenitized at 1410 Degrees Fahr., Transformed at 1350 Degrees Fahr. for 3 Hours, Water-Quenched and Tempered. Fig. 28—Edge of Sample. $\times 100$. Fig. 29—Core, 0.023 Inch From Edge. $\times 1000$. Fig. 30—Medium Carbon Zone, 0.011 Inch From Edge. $\times 1000$. Fig. 31—Low Carbon Zone, 0.008 Inch From Edge. Nital Etch. $\times 1000$.

region, the latter being completely transformed, whereas the lower carbon zones were only 30 to 50 per cent transformed, as in Figs. 28, 29, 30 and 31. The results of this experiment showed that steels of all carbon contents could be transformed to a spheroidal product under carefully controlled temperature conditions if sufficient time were allowed for the transformation to be completed in the lower carbon regions.

SUMMARY OF RESULTS OF LABORATORY WORK

It has now been shown that austenite of plain carbon steels can be made to transform directly to spheroidite, rather than to pearlite. To produce this transformation product, it is necessary that the austenite be heated to a temperature not far above its critical, and transformed at a temperature not far below its critical. The carbon content of the steel controls the temperatures within which the austenite must be heated, and transformed, for the production of a spheroidal product; the hypereutectoid steels allow most deviation from the optimum conditions, and the intermediate carbon steels, say 0.30 to 0.70 per cent carbon, least.

Unfortunately, the authors are not yet prepared to offer a reasonable explanation for the behavior of the austenite, which under one set of conditions transforms to spheroidite, and under other conditions not much different, transforms to pearlite. Although it appears that the austenite associated with excess carbides transforms to spheroidite most readily, nevertheless it was found that in some cases austenite which appeared to have all the carbon in solution transformed to a spheroidal product, and in other cases a lamellar product was formed from austenite which did not have all the carbon in solid solution. The discovery of the exact mechanism by which the transformation is controlled will require considerable exploration beyond the facts which the authors have disclosed, and it is hoped that other investigators will be encouraged to work in this field.

APPLICATIONS OF ISOTHERMAL TRANSFORMATION TO PRACTICAL ANNEALING OPERATIONS

The relatively short procedures given in Table III for producing spheroidal structures in plain carbon steels are somewhat idealized in that they do not allow for the time required to heat the steel to the preheating, or austenitizing, temperatures. Furthermore, relatively large loads in batch type furnaces may heat sufficiently slowly

on the way up to the austenitizing temperature to make it unnecessary to hold the temperature purposely in the subcritical region. Also, the cooling rate of large loads in well insulated furnaces may be sufficiently slow to allow transformation to be completed at the high temperatures in steels like D and E, and thus eliminate the need for holding the temperature at some level during the cooling. Under such conditions, the steel can be completely spheroidized by a furnace cool at the "natural" rate of the loaded furnace, and the load can be "pulled" when the steel temperature has dropped to about 1250 degrees Fahr. (675 degrees Cent.), and allowed to finish cooling in air. For steels like A, B and C, which have relatively slow transformation rates, very slow cooling from the austenitizing temperature down to the transformation temperature is not desirable because the time spent in cooling down to the holding level is more or less wasted, and would be much better used at the holding temperature. Under any circumstances, continued slow cooling after the steel temperature has dropped to about 1250 degrees Fahr. (675 degrees Cent.) is of no advantage. If the articles to be annealed are relatively small forgings which would change temperature relatively rapidly, the annealing for a spheroidal structure can quite satisfactorily be carried out in a continuous operation.

REVIEW OF OLDER LITERATURE ON SPHEROIDIZATION

Having arrived at a practical method for spheroidizing steels, which differed from the procedures given in standard books of reference, the authors felt it incumbent on them to review the older literature to see whether others had published observations which were in any way similar to theirs. The earliest work on spheroidizing is reviewed in two delightfully written papers by Howe and Levy (4a, 4b). These writers held the classic view that austenite transforms to sorbite, then to pearlite, and that with prolonged holding at temperatures below the critical, the pearlite "divorces" to ferrite and spheroids of carbide. They originated the expression, "divorcing of pearlite" and, in justification of it, they stated, "After all, it is pretty pat, suggesting that a pair well mated, with regularity and formality, parts company." Further, "The divorce of pearlite is very capricious. Masses of sharply lamellar pearlite may lie very near regions in which not only is the divorce of pearlite complete but in addition, its divorced elements have coalesced into large masses. . . . The pearlite masses in the same section differ greatly

among themselves in their attachment, some parting company at the first cross word, and others bearing and forebearing, patiently." They worked with steels of a wide range of carbon, but they heated all of them at 1650 degrees Fahr. (900 degrees Cent.), and it is clear now that they should have had difficulty in getting complete spheroidization. They did state that very high heating, at 2190 to 2370 degrees Fahr. (1200 to 1300 degrees Cent.) prior to cooling, as well as long time holding at lower heating temperatures, retarded spheroidization.

At about the time that Howe and Levy at Columbia University were doing the work just referred to, Hanemann and Morawe in Germany (4c) were making somewhat similar observations. They too heated their steels at 1650 degrees Fahr. (900 degrees Cent.) and experimented with various rates of cooling to temperatures below A_{r1} , to produce spheroidal structures, and they found that different steels required different rates of cooling.

Portevin and Bernard (4d, 4e), in France, reported in 1915 that they did not find much spheroidal product formed during cooling from 1470 degrees Fahr. (800 degrees Cent.) and holding at 1290 degrees Fahr. (700 degrees Cent.). However, in 1921, they reported that slow cooling favored coalescence.

In Japan, Honda and Saito (4f) worked with a number of steels varying from 0.31 per cent carbon to 1.59 per cent carbon. They heated these to various temperatures from about 1255 to 1650 degrees Fahr. (680 to 900 degrees Cent.), held them just twenty minutes at temperature, and then slowly cooled them. They found that all steels would develop spheroids on slow cooling if the heating temperatures were between about 1346 and 1355 degrees Fahr. (730 and 735 degrees Cent.); that if the heating temperature was about 1380 degrees Fahr. (750 degrees Cent.), the lower carbon steels—0.80 per cent and under—would have a mixed structure of pearlite and spheroids, whereas the higher carbon steels would still be spheroidal; and finally, if the heating temperature went up to 1650 degrees Fahr. (900 degrees Cent.), all steels would contain pearlite on slow cooling. It is to be assumed from a more recent statement by Honda (4g) that these authors believed the spheroidal product to be the last step in a transition, during the very slow cooling, from austenite, to martensite, to pearlite. It is interesting to note that their work did not find favor with English metallurgists, since F. C. Thompson, in a discussion three years later, stated, "This paper (by Desch

and Roberts), (4h), would deserve a hearty welcome if it did no more than give a death blow to the Japanese heresy that the production of spheroidal pearlite depended on taking the material above the A_1 point."

Although the very earliest observations on spheroidization had been made in England by Arnold (4i), and by Stead (4j), not much additional information was published there until J. H. Whitley (4k, 4l), in 1918, showed that pearlite, which had been cold-worked, spheroidized much more readily at 1202 to 1256 degrees Fahr. (650 to 680 degrees Cent.) than the pearlite in the "as-rolled" condition of a hot-rolled bar. Four years later, this same author while attempting to establish accurately the A_{r1} point of carbon steels, observed that he obtained spheroids when his steels were cooled to 1305 degrees Fahr. (710 degrees Cent.) and 1300 degrees Fahr. (705 degrees Cent.), providing his heating times were very short—three to five minutes—and his heating temperatures were not much over the critical. He did not obtain completely spheroidal structures. More recently in England, Carpenter and Robertson (4m) studied the transformations in carbon steels during slow cooling and they showed that the structure formed from a relatively low heating temperature was mainly spheroidal, whereas that formed from a high heating temperature was lamellar.

There have been a few papers on spheroidizing in this country since the publication of the papers by Howe and Levy. Emmons (4n) showed that the relative amounts of lamellar and spheroidal pearlite in annealed tool steel could be altered by the rates at which the steel was allowed to cool from above the critical. MacPherran and Harper (4o) discussed the "loneal" treatment for producing a spheroidized structure in steels—similar to Steel B of this paper, the treatment consisting of very long heatings, up to 72 hours at about 1250 degrees Fahr. (675 degrees Cent.). Foley, Clayton and Remmers (4p) discussed the effect of cooling rates from 1380 degrees Fahr. on the properties of a 0.75 per cent carbon, 0.52 per cent manganese steel. They found that the slower the cooling, the more spheroidal the product, but they were of the opinion that pearlite formed first and then spheroidized during the retarded cooling.

Thus, after a reasonably diligent search, the authors have found no suggestion, in the literature, of a practical procedure for spheroidizing steels other than the traditional ones involving very slow cooling, or very long subcritical heating, or both. It may be that

this results from the fact that the practical man has relied on his furnace control thermocouples to indicate the steel temperature, whereas the present authors have found that the steel temperature may be 20 to 50 degrees Fahr. (11 to 28 degrees Cent.) different from the temperature indicated by the furnace thermocouples, especially if the steel is packed in pipes or boxes. Such inaccuracies in temperature could not be tolerated in some of the more efficient annealing cycles based on isothermal transformation. However, experience with many tons of steel annealed on a commercial scale has proved that this principle of annealing is altogether practical.

ACKNOWLEDGMENTS

The authors are grateful to the executives of the Crucible Steel Co. of America for permission to publish the data obtained at the Research Laboratory. Mr. A. A. Hassan, Jr., chief metallurgist of the Atha Works, was of great assistance in applying the laboratory findings to the development of improved commercial annealing operations. Dr. J. Z. Briggs very kindly assisted with the bibliography.

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DISCUSSION

Written Discussion: By W. R. Frazer, Union Twist Drill Co., Athol, Mass.

This excellent paper adds another important chapter to the various studies that have been made of isothermal transformation that will take place in steel under certain conditions. These authors have confined their investigation to the upper part of the S curve discovered by Davenport and Bain in 1930. They have explored that portion of the curve just under the austenitizing temperature and find that austenite transforms to a spheroidal carbide in this zone if the steel is austenitized at just over the critical. This is of great commercial value due to the fact that current spheroidizing annealing cycles take a very long time and the new cycle disclosed in this paper is comparatively short, which will effect definite economy in both time and cost of operation.

It is noted that the small test-pieces were austenitized at temperature for one hour before being quenched to the transformation temperature. This seems like an unusually long time to hold such a small piece at temperature and we would like to ask the authors if this was the shortest austenitizing time they found which entirely eliminated the lamellar structure in the original material.

If we compare the annealing procedure recommended in Table III with the transformation temperature-time curves in Fig. 15, we observe that a much longer transformation temperature holding time is used commercially than the curves show to be necessary for complete transformation. For example, the curve for Steel E (1.16 carbon) shows complete transformation after holding 12 minutes at 1335 degrees Fahr. whereas the recommended annealing cycle is to hold at 1335 degrees Fahr. for 3 hours. May we ask if this additional transformation time is recommended for commercial annealing because of the extremely accurate steel temperature control necessary to obtain satisfactory results? Also in Table III we note that the lower carbon steels which are more difficult to spheroidize are "preheated" at just under the austenitizing temperature. Does not this "preheat" coalesce the carbides of the original lamellar structure and start the spheroidization that takes place as the process proceeds?

The authors are to be congratulated on this timely investigation which gives us a much quicker method of spheroidizing carbon steel.

Written Discussion: By John J. B. Rutherford, research metallurgist, The Babcock & Wilcox Tube Company, Beaver Falls, Pa.

According to the definitions contained in the A.S.M. Metals Handbook, 1939 Edition, Full-annealing and Spheroidizing may be synonymous when

applied to tool steels but in the same handbook, it may be noted, there is no distinctive definition of "tool steel." There may be some advantage in qualifying "tool steels" as "alloy steels" since the process described as Spheroidizing for tool steels has also been applied commercially to various other types of S.A.E. alloy steels which do not fall in the category of tool steels. The present paper clearly demonstrates this fact and, in an excellent manner, subscribes further to the important phase of heat treating which is intermediate between Process- and Full-annealing.

The keynote of this process of rapid spheroidization is struck when the authors describe their attempts to establish the minimum austenitizing temperatures at which they observed *austenite plus nonlamellar carbides*; this initial condition is necessary for the process described regardless of the subsequent rate of cooling. The authors, by a more critical analysis of their data, might have established certain fundamental concepts, dangerous as this may seem. For simple carbon steels, the austenitizing temperatures depend on extra-equilibrium considerations but in tool steels, in the presence of carbide-forming elements, three phases may exist at equilibrium, viz.: ferrite, austenite and Fe_3X_m , C_n or XC.

If it is considered, and in the writer's opinion experience bears this out, that some discrete form of nucleation such as undissolved carbides, inclusions, etc., must be present in the austenite before spheroidization occurs directly on transformation, then many common observations can be readily explained. The authors have found the *minimum* austenitizing temperature most efficacious in spheroidizing and, in the case of the lower carbon steels, found a prior treatment to coalesce the carbide (2 hours at 1320 to 1340 degrees Fahr.) advisable. "It was found that for steels A, B and C, the amount of lamellar pearlite in the transformation product increased at once as the austenitizing temperature was raised," and the same applies to the other steels with this difference, the degree of overheating necessary to destroy nucleation is varied in magnitude. The presence of lamellar pearlite in the decarburized rim of a spheroidized bar, a common occurrence in practice, seems logically related to the fact that the optimum austenitizing temperature for the body of the bar will *completely* dissolve the carbon at the rim and therefore destroy the nucleation required.

The heat treatment described by the authors has been practiced commercially with a greater latitude of temperature control than is indicated from laboratory experiments. In the case of S.A.E. 52100 steel, for example, the usual laboratory technique of heating a small sample to the desired temperature within some 15 minutes requires an austenitizing temperature limited to about a 50-degree Fahr. range (prior structure; lamellar pearlite) in order to obtain complete spheroidization whereas commercial practice requiring in the order of 5 to 20 hours heating establishes a much wider range for satisfactory austenitizing. It may be in order to point out that large furnace charges of alloy steels are annealed slowly not alone for strictly metallurgical reasons but primarily to permit temperature equalization; this time factor has been reduced very considerably, in many cases, by the use of suitable equipment for continuous annealing.

The writer proposes the following concepts for the authors' consideration:

1. The formation of lamellar pearlite is inherently associated with the "unrestricted" growth, from austenite, of the orthorhombic Fe_3C type of carbide.

2. Under certain conditions of nucleation, spheroidized carbide will result from the transformation of austenite instead of lamellar pearlite. Higher transformation temperatures permit of better opportunity for the carbide to coalesce around these nuclei.

3. In steels containing large quantities of carbide-forming, alloying elements, transformation of austenite at the higher temperatures customarily obtains irregular polygonal carbide constituents in place of lamellar pearlite; this is probably associated with the modification of the carbide crystal structure.

Oral Discussion

B. R. QUENEAU:¹ Davenport and Bain have shown that a homogeneous austenite will transform isothermally at temperatures just below the equilibrium critical range to a product that is lamellar in structure. It should be emphasized that they made an effort to have the austenite homogeneous, with all carbides in solution and therefore relatively high austenitizing temperatures were used.

According to my belief when austenite containing undissolved carbides transforms close to the critical there is a tendency for the carbide to precipitate out on the spheroids already present, and thus a nonlamellar structure is produced.

It may well be that, in addition to carbides, there are other particles present after low temperature austenitizing that would act as nuclei for carbide growth. However, the authors have not shown conclusively in their experiments with the decarburized steels, that some carbides were not always present in the austenite from which the spheroidal structures were formed.

The absence of lamellae in the transformation product at 1350 degrees Fahr. may be attributed to a slow rate of reaction which allows the carbon to diffuse to existing nuclei, few though they may be. At 1300 degrees Fahr. where the rate of reaction is faster, the carbon does not have sufficient time to diffuse and the carbide is rejected in the form of lamellae.

This same explanation may account for the general observations of the authors that lower carbon steels, which naturally have fewer undissolved carbides after austenitizing, are more difficult to spheroidize and that a long time at low austenitizing temperatures has in general the same effect as a short time at high austenitizing temperatures.

Authors' Reply

The authors appreciate the discussions by Messrs. Frazer, Rutherford, and Queneau.

In connection with Dr. Frazer's questions, it should have been emphasized that the reaction—ferrite plus lamellar carbides, to austenite plus spheroidal carbides—is a sluggish one at the low austenitizing temperatures, and an appreciable time was required, as shown by examination of microsections of samples

¹Instructor of metallography and ferrous metallurgy, Columbia University, New York.

held for successively longer times, to complete this reaction. The time required at say 1365 degrees Fahr. (740 degrees Cent.) would be considerably longer than that required at 1380 degrees Fahr. (749 degrees Cent.), and a compromise was made by selecting "minimum austenitizing temperatures" at which the reaction would be complete in one hour, or less. Dr. Frazer is quite correct in assuming that the transformation time recommended for commercial annealing includes an appreciable safety factor to allow for inaccuracies in furnace temperature, and differences which may exist in reaction rates of commercial steels. Furthermore the transformation curves were based on materials which had been preheated before being austenitized, and this pretreatment speeds up the reaction.

In their efforts to account for the formation of the spheroidal carbide as a direct product of the transformation of austenite, the authors have also considered, as have the discussers, the probability that residual carbides are necessary to act as nuclei for the formation of new spheroids. However, it is submitted that in the sample of S.A.E.-4345, austenitized at 1450 degrees Fahr. (790 degrees Cent.), Fig. 1; in the sample of the 0.46 per cent carbon steel B, austenitized at 1370 degrees Fahr., Figs. 12, 16 and 17; and in the lower carbon regions of the decarburized rim of the hypereutectoid steel heated at 1410 degrees Fahr. (765 degrees Cent.), Figs. 25 to 31, there is no evidence at 1000 magnification of the presence of residual carbides in the austenite, and yet spheroidal products were formed. Furthermore Steel D, 0.85 per cent carbon, heated to 1470 degrees Fahr. (800 degrees Cent.), showed no residual carbides in the austenite, and yet when this austenite transformed at 1340 degrees Fahr. (725 degrees Cent.) at least 50 per cent of the product was spheroidal. If the mere absence of residual carbides is the requirement for the formation of a lamellar product, then all the specimens referred to should have contained only lamellar carbides. The authors are inclined to agree with Professor Queneau that "there are other particles (not carbide) present after low temperature austenitizing that would act as nuclei for (spheroidal) carbide growth". However, before such a hypothesis can gain any credence, the identity of such particles must be established, as well as the solubility of such particles in both austenite and ferrite.

ANNEALING OF LOW CARBON RIMMED, ALUMINUM-KILLED AND SILICON-KILLED STEELS

BY W. P. WALLACE AND R. L. RICKETT

Abstract

The effect of time and temperature of annealing and amount of prior cold reduction on the structure and hardness of three commercial low carbon steels when annealed under carefully controlled conditions was investigated. A rimmed, a silicon-killed and an aluminum-killed steel, all containing 0.15 per cent carbon or less, and all cold reduced various amounts ranging up to 81 per cent reduction in area were employed. The temperature range covered was 900 to 1350 degrees Fahr. and periods of 15 minutes, 2 hours and 16 hours at each annealing temperature were used.

Results of the investigation are expressed in the form of numerous curves relating the annealing variables, amount of prior cold reduction, and type of steel, severally or individually, to hardness and in some instances degree of recrystallization and grain size. Photomicrographs of a few representative structures are included.

THE recrystallization of cold-worked metals has been extensively investigated and a large number of publications dealing with this subject may be found in the technical literature. The amount of published information dealing with the influence of systematic variation in annealing conditions on the structure and properties of cold-rolled or cold drawn low carbon steel as commercially produced is, however, rather limited; it was to obtain such information that the work to be described was undertaken.

The materials investigated were a rimmed, a silicon-killed and an aluminum-killed steel, all commercial open-hearth steels containing 0.15 per cent carbon or less. They were cold drawn various amounts up to 81 per cent reduction in area and annealed under closely controlled conditions at 900, 1000, 1100, 1200 or 1350 degrees Fahr. for 15 minutes, 2 hours or 16 hours.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, W. P. Wallace is associated with the Columbia Steel Co., Torrance, California, and R. L. Rickett is associated with the Research Laboratory, United States Steel Corporation, Kearny, N. J. Manuscript received June 26, 1939.

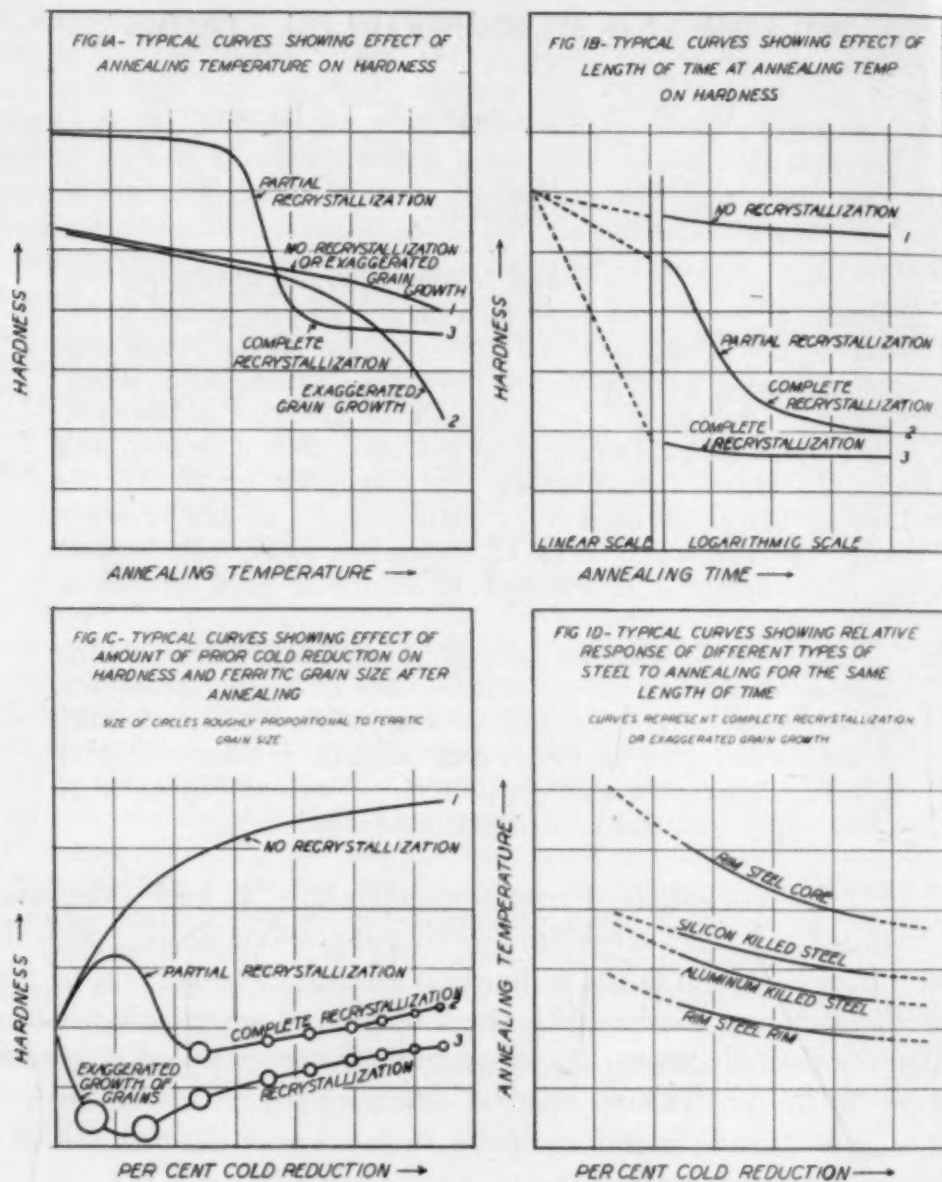


Fig. 1—Typical Curves Showing Effect of Annealing and Cold Reduction on Hardness and Grain Size.

The results obtained are shown more concisely by the curves of Figs. 3 to 12 than they can be expressed in words. The general trends observed are described below with the aid of typical curves as drawn in Fig. 1.

Effect of Annealing Temperature on Hardness and Grain Size—The effect of annealing temperature on hardness may be represented by curves of the type shown in Fig. 1A:

Curve 1 is characteristic of relatively small amounts of cold reduction and short annealing time; there is a slight "recovery" or de-

crease in hardness after annealing at the higher temperatures but no observable change in microstructure.

Curve 2 is characteristic of slightly greater amounts of cold reduction or, in some instances, of longer annealing times than in Curve 1; marked softening occurs at the higher annealing temperatures and is accompanied by the formation of very coarse ferrite grains.

Curve 3 is typical of still greater amounts of prior cold reduction. Slight softening, or "recovery," with no observable change in microstructure occurs on annealing at the lower temperatures; marked softening together with recrystallization and grain growth occurs over an intermediate temperature range, and a slight further softening and continued grain growth takes place on annealing at higher temperatures.

Effect of Time of Annealing on Hardness and Grain Size—Hardness results may be represented by curves of the type shown in Fig. 1B:

Curve 1 is characteristic of annealing temperatures and amounts of prior cold reduction that do not result in "exaggerated" grain growth or recrystallization.

Curve 2 shows slight softening after a short time, and marked softening, resulting from recrystallization or grain growth, after a longer time at temperature; this type of curve is characteristic of certain combinations of temperature and amount of cold reduction.

Curve 3 is the type observed when complete recrystallization occurs almost immediately, as for instance on annealing at the higher temperatures after a considerable amount of cold reduction; longer times decrease hardness only slightly and tend to increase ferrite grain size.

Effect of Amount of Prior Cold Reduction on Hardness and Grain Size After Annealing—Typical curves are shown in Fig. 1C:

Curve 1 represents the result of annealing at temperatures that do not cause "exaggerated" grain growth or recrystallization; hardness increases with increasing amounts of prior cold deformation.

Curve 2 is typical of instances where intermediate amounts of prior cold reduction result in partial recrystallization, and higher amounts, in complete recrystallization; hardness increases and grain size decreases as the severity of deformation becomes greater than that necessary to cause complete recrystallization.

Curve 3 indicates a minimum hardness and very coarse ferrite

grain size in material previously cold reduced a relatively small amount; for higher reductions, hardness increases and grain size decreases.

Effect of Type of Steel on Annealing Behavior—Fig. 1D shows the comparative response to annealing, for a given period, of the different types of steel investigated in terms of the minimum annealing temperature and amount of prior cold reduction necessary to produce essentially complete recrystallization or exaggerated grain growth, as the case may be. Of the materials investigated, the rim portion of the rimmed steel is affected most readily and the core portion of the rimmed steel least readily, while the silicon- and aluminum-killed steels are affected very much alike and in an intermediate manner.

Much more pronounced grain growth after small amounts of deformation occurs in the rim of the rimmed steel than in any of the other materials. In the completely recrystallized condition and for similar amounts of cold reduction and annealing, the rimmed steel rim is softest and coarsest-grained of the materials investigated; the two killed steels and the core of the rimmed steel are much alike in hardness and grain size.

These conclusions as to the effect of type of steel on annealing behavior may apply only to the particular materials investigated.

Effect of Annealing on Microstructure—The silicon-killed and aluminum-killed steels, before annealing, contained pearlite areas that were elongated and broken up to an extent depending on the amount of prior cold reduction. The carbides in these cold-worked pearlitic areas were only slightly affected by annealing at the lower temperatures but were definitely spheroidized when heated to higher temperatures, providing the A_1 temperature was not exceeded; increased time at these higher temperatures increased the amount of spheroidization. The rimmed steel contained carbide particles rather than pearlite, and these carbide particles were broken up somewhat by considerable amounts of cold drawing. No marked change occurred in these carbide particles during annealing except in the material annealed at 1350 degrees Fahr.; evidently at this temperature the carbon present was essentially all in solution and on subsequent cooling irregular, intergranular carbide particles formed.

In general, it may be said that on annealing at temperatures in range 1200 to 1350 degrees Fahr. only slight changes in hardness and grain size occurred after the first two hours and, in some in-

stances, even after the first 15 minutes at the annealing temperature. It appears probable that with respect to hardness and grain size no very useful purpose will be served by annealing these types of steel for longer than two hours at temperatures in this range; nevertheless, the degree of spheroidization, the distribution of carbide particles, and properties other than hardness may be altered by annealing for longer than two hours, providing the A_1 temperature is not exceeded.

MATERIAL AND PROCEDURE

Composition of the material as determined by analysis of the actual bars used is as follows:

	C	Mn	P	S	Si	Ferrite Grain Size (A.S.T.M.) (Hot-Rolled)
Aluminum-killed	0.15	0.44	0.019	0.049	0.040	7
Silicon-killed (Rephosphorized)	0.11	0.31	0.051	0.054	0.095	7
Rimmed steel core ...	0.03	0.09	0.007	0.062	0.006	6
Rimmed steel rim ...	0.02	0.08	0.002	0.016	0.002	6

These steels were hot-rolled into 1-inch rounds and then cold reduced as follows:

Draft	Diameter of Finished Section Inches	Per Cent Reduction in Area
0 (Hot-Rolled)	1.003	0
1	0.968	7
2	0.906	18
3	0.842	30
4	0.749	44
5	0.689	53
6	0.622	61
7	0.561	69
8	0.498	75
9	0.433	81

Annealing was carried out at temperatures of 900, 1000, 1100, 1200 or 1350 degrees Fahr: for periods of 15 minutes, 2 hours or 16 hours at each temperature. A group of samples consisting of one from each steel for each amount of cold reduction listed above, or 30 samples in all, was annealed for each period of time at each temperature. The samples were $\frac{1}{2}$ inch long and were packed in cast iron chips to minimize scaling and decarburization, in a sheet metal container sealed with refractory cement. A thermocouple inserted into the center of the box and connected to a recorder was used to measure the temperature of the specimens. The period at the annealing temperature was arbitrarily considered as beginning

when the indicated temperature reached within 10 degrees Fahr. of the desired value; approximately 1.5 hours were required for the specimens to reach this temperature. At the completion of the run, the box was removed from the furnace and allowed to cool in air, requiring about two hours to cool down to 150 degrees Fahr. Fig. 2 shows typical heating and cooling curves. The specimens were ground down somewhat, finishing on 320 paper, and hardness readings made on the cross section, except for the rimmed

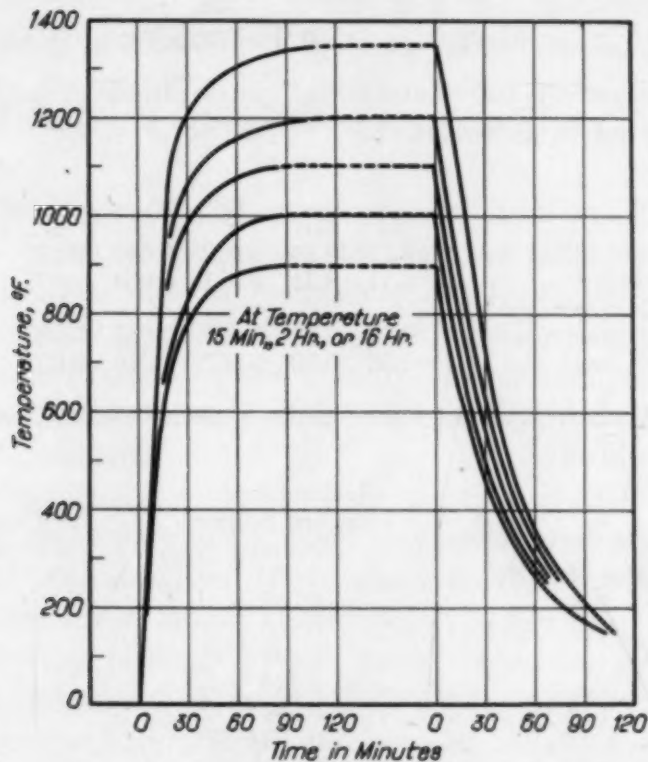


Fig. 2—Typical Heating and Cooling Curves.

steel rim in which case the hardness determinations were made on a longitudinal section. A preliminary investigation showed that the hardness was the same whether measured on a longitudinal or on a cross section, except, of course, for the rimmed steel. Longitudinal sections near the center of the specimen were used for microscopic examination.

RESULTS

Microstructure—The structural changes that may occur during annealing at temperatures within the range used in this investigation are illustrated in Figs. 3 and 4. Fig. 3A shows a typical cold-worked structure, in this case after 61 per cent reduction in area

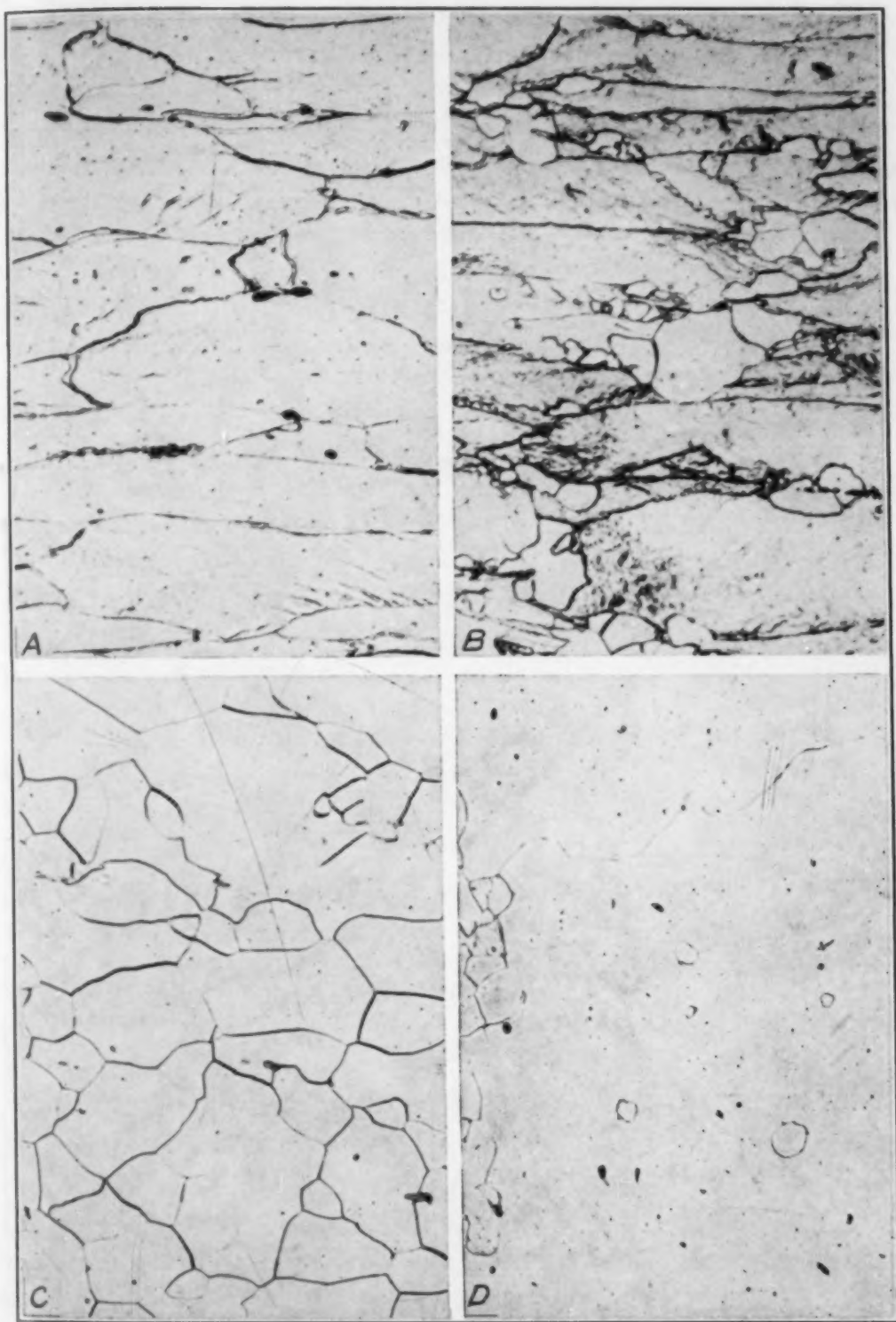


Fig. 3—Typical Ferrite Microstructures. Fig. 3A—Cold Drawn Structure. Rim of Rimmed Steel After 61 Per Cent Reduction. $\times 500$. Fig. 3B—Partially Recrystallized Structure. Same Material as 3A. Annealed 15 Minutes at 1000 Degrees Fahr. $\times 500$. Fig. 3C—Completely Recrystallized Ferrite. Same Material as 3A and B. Annealed 16 Hours at 1000 Degrees Fahr. $\times 500$. Fig. 3D—"Exaggerated" ("Germinative") Grain Growth in Rim of Rimmed Steel Reduced 7 Per Cent and Annealed 2 Hours at 1350 Degrees Fahr. $\times 100$.

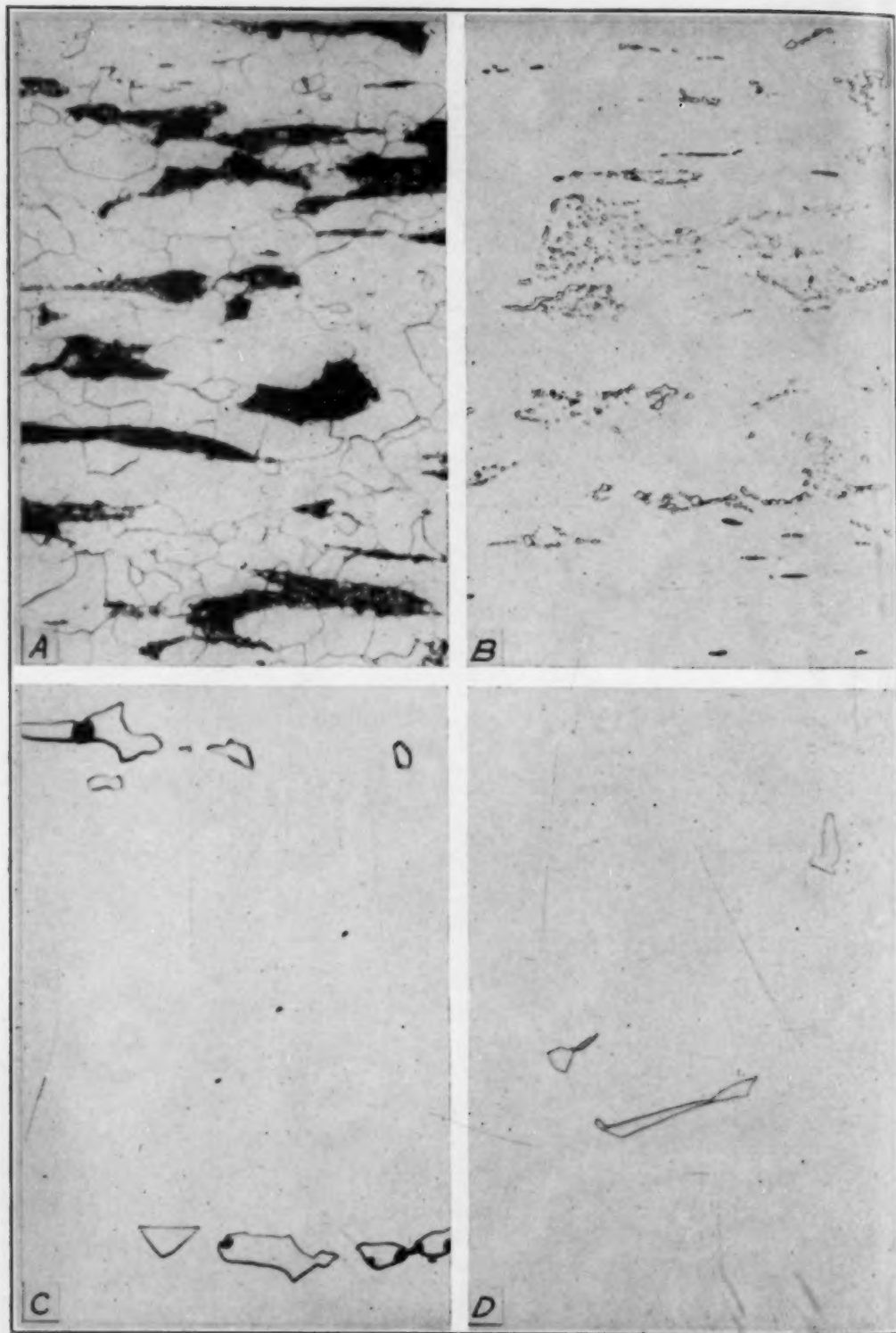


Fig. 4—Typical Carbide Structures. Fig. 4A—Complete Recrystallization of Ferrite and Partial Spheroidization of Carbides. Al-killed Steel of 61 Per Cent Reduction. Annealed 2 Hours at 1100 Degrees Fahr. $\times 500$. Fig. 4B—More Complete Spheroidization of Carbides. Same Material as 4A. Annealed 16 Hours at 1200 Degrees Fahr. $\times 500$. Fig. 4C—Carbides Typical of Those in Rimmed Steel as Cold Drawn Material and as Annealed at 1200 Degrees Fahr. or Below. $\times 1000$. Fig. 4D—Structure Resulting from Annealing Rimmed Steel at 1350 Degrees Fahr. Carbides Precipitated in Ferrite Grain Boundaries. $\times 1000$.

by cold drawing; higher reductions cause still more distortion of the grains while lower reductions result in less distortion, as would be expected. The lowest amount of cold reduction which produced definite deformation of the grains observable under the microscope was 30 per cent. Fig. 3B illustrates partial recrystallization on annealing; the new, unstrained grains appear to form mostly in old grain boundaries and are very small at first. Continued formation of new grains together with some grain growth results in a completely recrystallized and generally equiaxed structure, as shown in Fig. 3C; further heating may produce a small additional increase in grain size. This mode of recrystallization and grain growth represents the changes in the ferrite grains observed in this investigation except when the amount of cold reduction prior to annealing was, with a few exceptions, 18 per cent or less. After these smaller amounts of cold reduction no distortion of the ferrite grains was evident, nor were small, new equiaxed grains ever observed after annealing; relatively coarse grains, as shown in Fig. 3D, were found after annealing at the higher temperatures, particularly after annealing for a comparatively long time at these temperatures. The formation of these relatively coarse grains on annealing will be referred to in this discussion as "exaggerated" or "germinative" grain growth rather than recrystallization.

Fig. 4 shows typical carbide distributions. In Fig. 4A the originally deformed pearlitic areas are still elongated even though the surrounding ferrite is completely recrystallized; some spheroidization of the carbide may be observed in Fig. 4A and more complete spheroidization in Fig. 4B, which represents the effect of a higher temperature and longer time of annealing. When annealed at 1350 degrees Fahr., some austenite formed in the aluminum- and silicon-killed steels used in this investigation; consequently, pearlite is present in the annealed product. In the very low carbon rimmed steel little or no austenite formed at 1350 degrees Fahr. or below; the material contains no pearlite after annealing but rather carbide in the form of discrete particles as shown in Figs. 4C and 4D.

Hardness and Grain Size—The effect of amount of cold reduction, annealing temperature, and annealing time on hardness and grain size is shown in Figs. 5 to 11. Fig. 5 is a photograph of a typical three-dimensional model showing the relationship of annealing temperature and amount of prior cold reduction to hardness and grain size. By projecting the curves representing the various

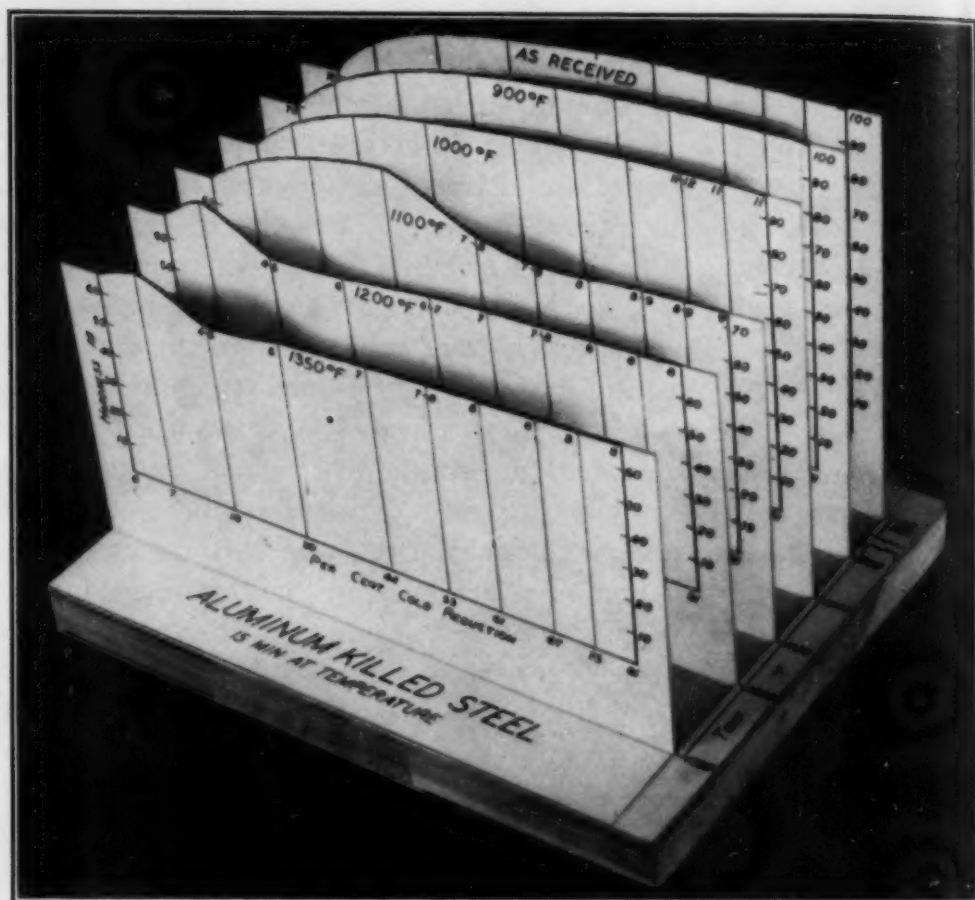


Fig. 5—Typical Three-Dimensional Diagram Showing Effect of Annealing Temperature and Amount of Cold Reduction on Hardness and Grain Size.

annealing temperatures as shown in Fig. 5 on a single plane, the charts of Figs. 6 to 9 inclusive were obtained. Fig. 10 shows the relationship between annealing time and hardness for three degrees of cold reduction. Fig. 11 shows the combinations of annealing temperature and amount of cold reduction corresponding to the beginning and end of recrystallization or exaggerated grain growth, or in other words, the range of most marked softening, for various lengths of time of annealing. Fig. 12 shows the effect of the type of steel on the annealing behavior. The effect of amount of prior cold reduction, temperature and time of annealing, and type of steel on hardness, grain size and amount of recrystallization will be discussed separately below.

Effect of Amount of Cold Reduction—The effect of amount of cold reduction on hardness for various annealing temperatures and lengths of time is shown in principle in Fig. 1C and in detail in Figs. 6 to 9 inclusive; these diagrams also show the degree of re-

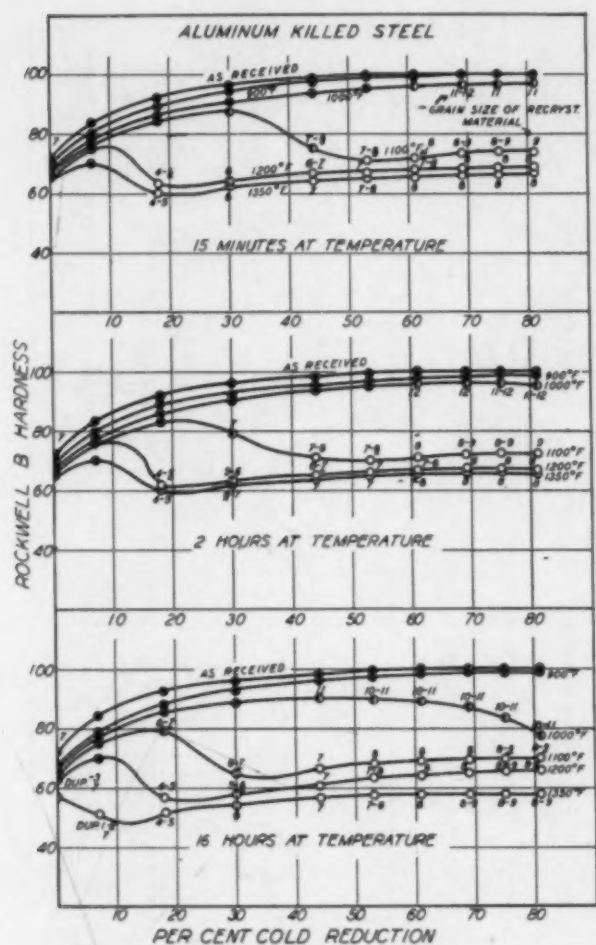


Fig. 6—Effect of Amount of Cold Reduction on Response to Annealing Treatment.

crystallization and the ferrite grain size of the recrystallized material. In the unannealed condition, hardness increases rapidly with the first few drafts and then more slowly as the amount of cold reduction increases. The hardness-per cent reduction curve remains similar in shape to that for the cold reduced, unannealed material when the annealing is carried out at the lower temperatures, the only change being a slight general decrease in hardness ("recovery"). The most severely cold-worked material recrystallizes most readily, while progressively higher temperatures or longer periods of time are required for recrystallization and softening as the amount of prior cold reduction decreases, as may be seen from Figs. 6 to 9 and Fig. 11. On annealing at comparatively high temperatures after relatively little cold reduction, in most cases not over 18 per cent, exaggerated grain growth occurs; Fig. 3D shows an example of the structure resulting from such exaggerated grain

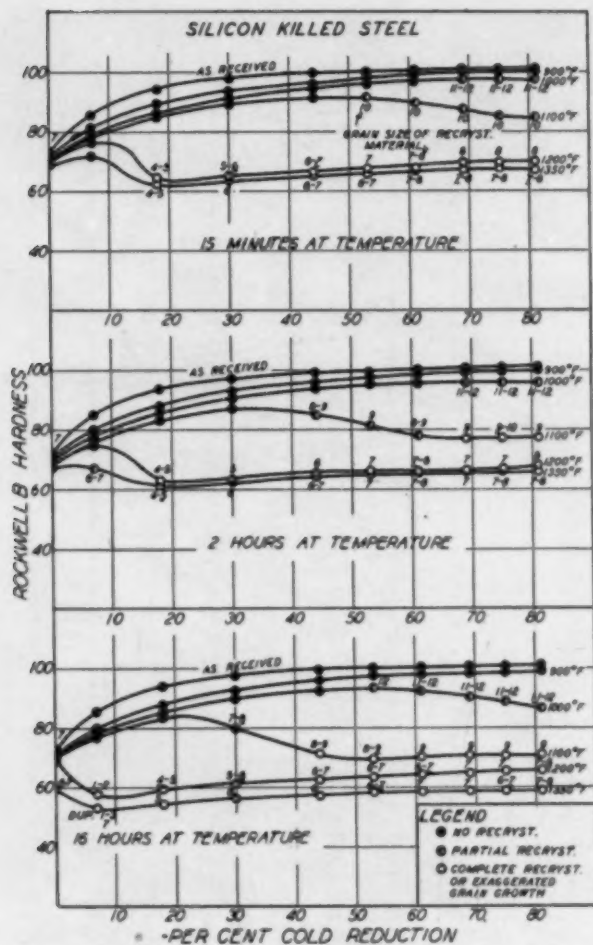


Fig. 7—Effect of Amount of Cold Reduction on Response to Annealing Treatment.

growth. Figs. 6 to 9 show that, for given annealing conditions resulting in exaggerated grain growth or complete recrystallization, hardness increases and grain size decreases, in general, as the amount of prior cold reduction increases.

Effect of Annealing Temperature—As annealing temperature increases, the hardness of the annealed product changes in the manner shown by the typical curves in Fig. 1A. Looking at Fig. 5, a set of similar curves can be imagined drawn parallel to the temperature axis of the model, each curve passing through the hardness values corresponding to a given amount of cold reduction but different annealing temperatures. Going now to the curves of Figs. 6 to 9, inclusive, the change in hardness with annealing temperature for each value of cold reduction can be seen by following along the corresponding vertical line in the diagram. Material cold reduced 7 per cent shows no pronounced change in hardness on annealing for

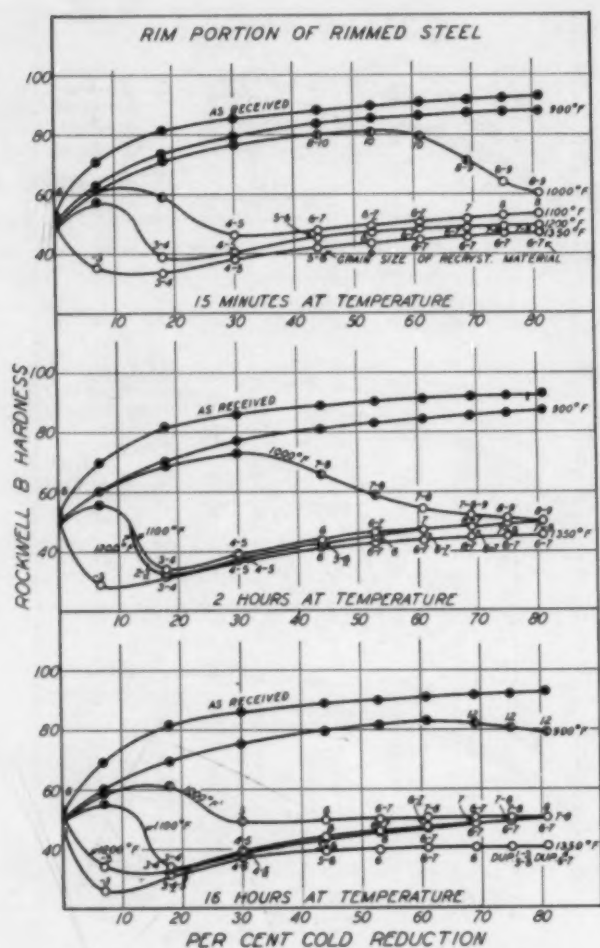


Fig. 8—Effect of Amount of Cold Reduction on Response to Annealing Treatment.

15 minutes in the temperature range up to 1200 degrees Fahr. With higher reductions or longer time at the annealing temperature, an abrupt drop in hardness occurs in a temperature range which depends upon the material being annealed, the amount of prior cold reduction and the length of time at the annealing temperature.

The gradual drop in hardness which results from annealing at the lower temperatures of the range investigated is not accompanied by any visible change in microstructure; this effect is usually termed "recovery." A marked drop in hardness, which is the result of recrystallization of the cold-worked metal or of exaggerated grain growth, occurs on annealing at somewhat higher temperatures; the minimum annealing temperature that results in such a sharp drop in hardness is lower the greater the amount of prior cold reduction, the longer the period of annealing, and varies with the steel being annealed. Temperatures above those necessary to

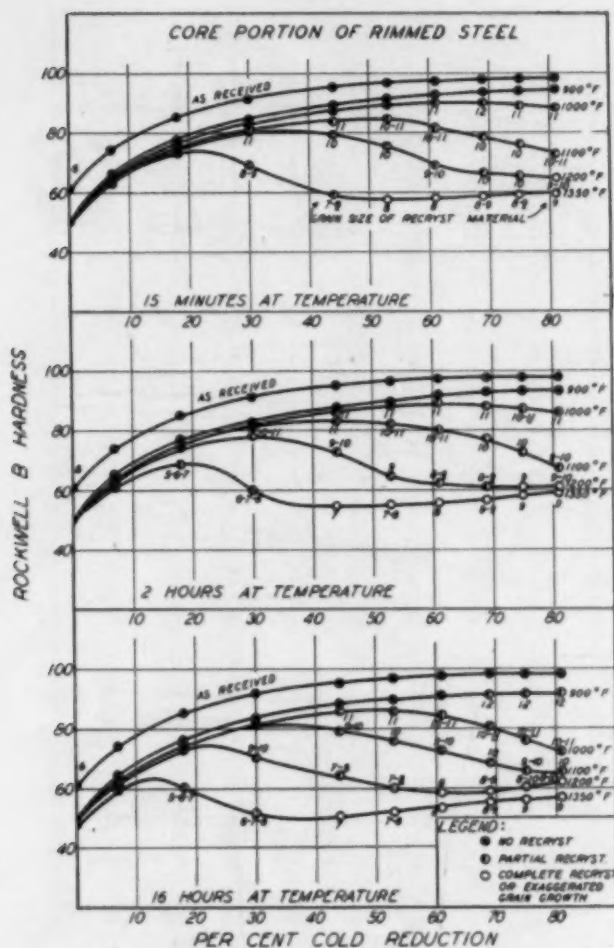


Fig. 9—Effect of Amount of Cold Reduction on Response to Annealing Treatment.

produce recrystallization result in a further slight decrease in hardness, and a tendency towards slightly coarser ferrite grain size. Fig. 11 shows that, in general, for a fixed annealing time, the higher the annealing temperature the narrower the range of cold reduction that results in only partial recrystallization.

Effect of Length of Time of Annealing—Typical curves showing the effect of length of time of annealing on hardness are presented in Fig. 1B, from which it may be observed that the effect of an increase in annealing time is quite similar to that of increasing annealing temperature (Fig. 1A). If the combination of temperature and amount of cold reduction is such that recrystallization or exaggerated grain growth does not occur, then the effect of increased time of annealing is to produce slight progressive softening of the material, Fig. 1B, Curve 1. For other combinations of temperature and prior reduction, recrystallization or exaggerated

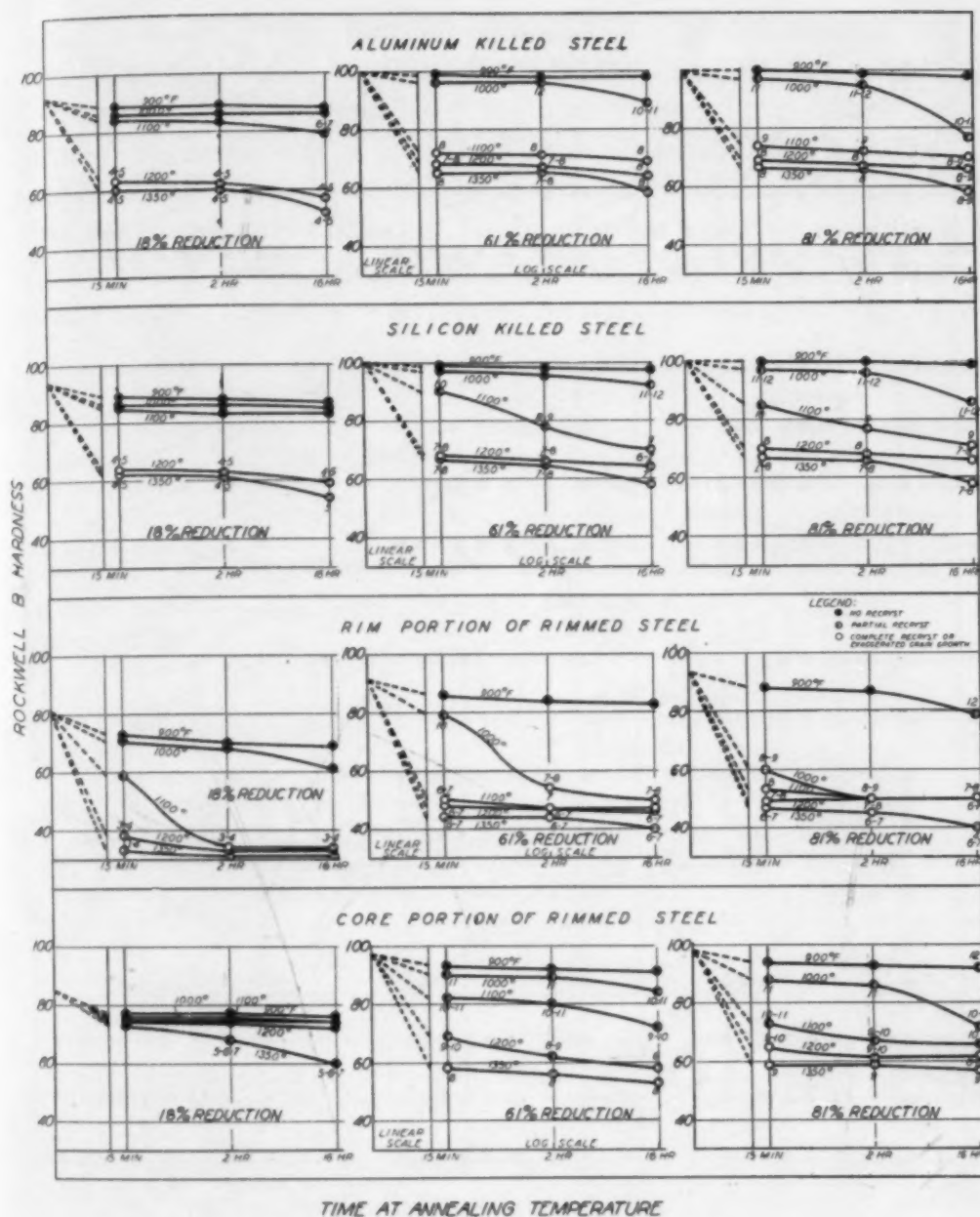


Fig. 10—Effect of Time at Annealing Temperature.

grain growth may not occur in a relatively short time, but will take place when the annealing time is increased; in such a case the hardness drops off slightly at first and then very markedly as the annealing time increases, Fig. 1B, Curve 2. For still other combinations of annealing temperature and cold reduction, recrystallization or grain growth and marked softening occur very quickly and increased annealing time causes only a slight further decrease in hardness and increase in grain size, Fig. 1B, Curve 3. Fig. 10

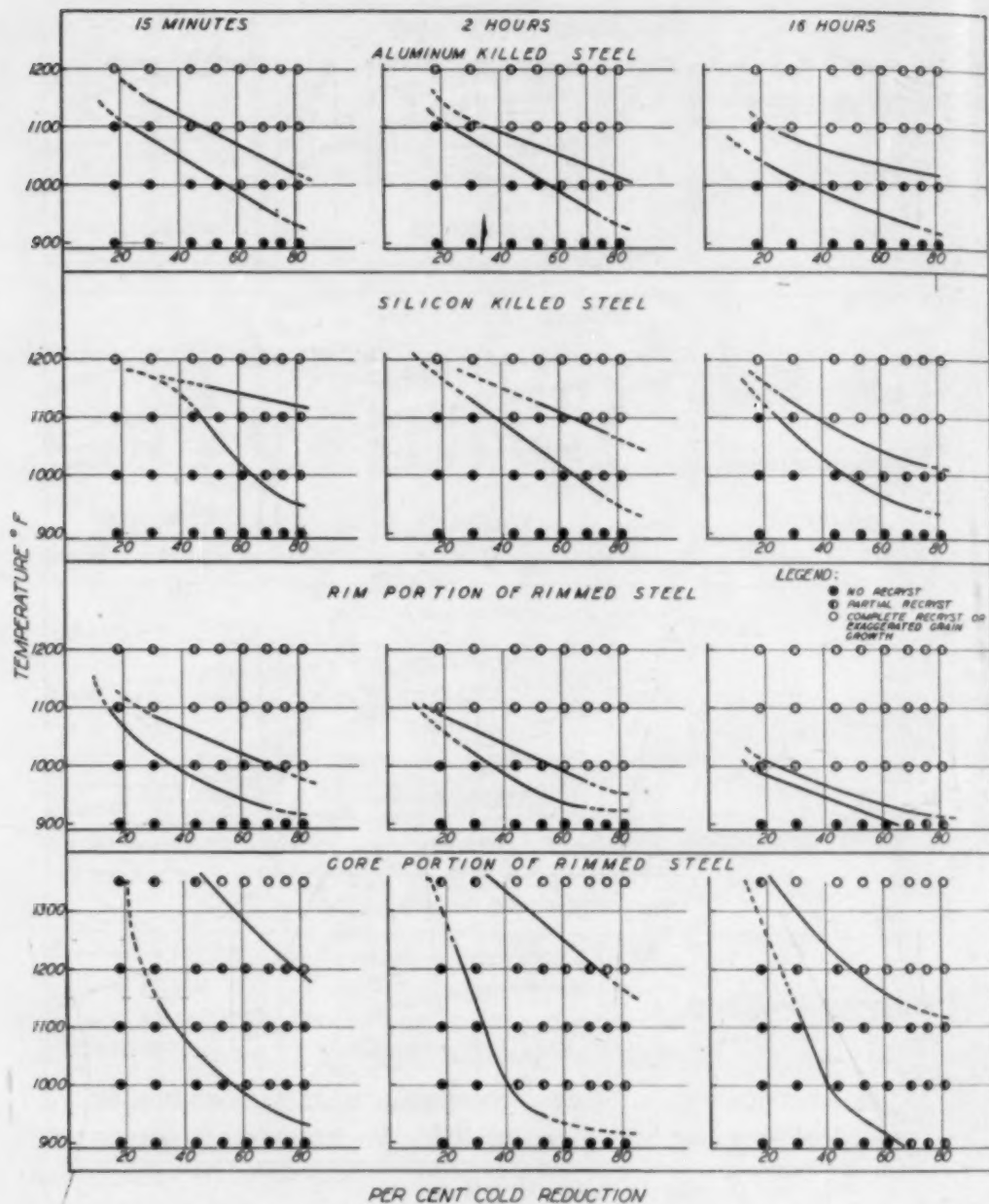


Fig. 11—Effect of Amount of Cold Reduction on Temperature Range of Recrystallization or Exaggerated Grain Growth.

shows curves of hardness versus annealing time for the materials included in this investigation after three different amounts of cold reduction, 18 per cent, 61 per cent, and 81 per cent.

Effect of Type of Steel—Fig. 11 shows that complete recrystallization or marked grain coarsening occurs more readily in the rim portions of the rimmed steel and less readily in the core portion of the rimmed steel than in the other two steels investigated. This difference in behavior is also brought out in Fig. 12, from

which it may be seen that for certain combinations of cold reduction, annealing temperature and annealing time, the rim portion of the rimmed steel recrystallizes completely and the core not at all. In such cases there is a very sharp line of demarcation between the recrystallized rim and the core; partial recrystallization occurs in the core immediately adjacent to the rim but none whatever a little deeper in the core. Another difference between rim and core of the rimmed steel, shown in Fig. 12, is the very coarse grain size in the

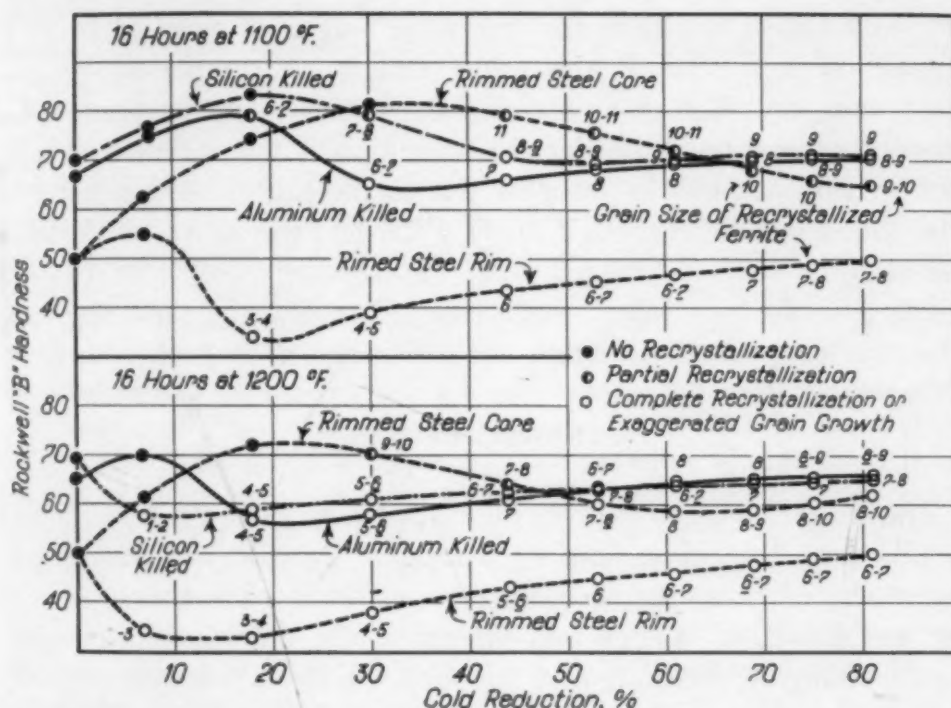


Fig. 12—Effect of Type of Steel and Amount of Cold Reduction on Response to Annealing at 1100 or 1200 Degrees Fahr. for 16 Hours.

rim as compared to the core when annealed at 1100 degrees Fahr. or higher for 16 hours after relatively small amounts of cold reduction. Fig. 11 shows a much wider range of conditions resulting in partial recrystallization of the core of the rimmed steel than for any of the other materials investigated. The curves of Figs. 6 to 9 inclusive indicate that the rim and core of the rimmed steel softened more on annealing at 900 degrees Fahr. ("recovered" more) than did the killed steels; this may be because 900 degrees Fahr. is nearer the temperature of beginning of recrystallization for the rimmed steel than for the other two. The difference in behavior between rim and core observed in this low carbon, low manganese steel may not be encountered to the same extent in annealing other

types of rimmed steel, but it is believed that the same qualitative differences will exist.

The aluminum-killed and silicon-killed steels investigated behave very much the same; both are less readily recrystallized than the rim portion of the rimmed steel and more readily than the core. The only consistent difference in behavior between the aluminum- and silicon-killed steels is observed when annealed at 1100 degrees Fahr. at which temperature the aluminum-killed steel recrystallizes and softens after smaller amounts of cold reduction than does the silicon-killed steel annealed for the same length of time (Fig. 12). With all other conditions the same, and when completely recrystallized, the two killed steels are slightly harder than the core of the rimmed steel and much harder than the rim of the rimmed steel as shown in Fig. 12; this is probably due to the higher carbon content of the killed steels. Although slightly harder, the killed steels tend to be slightly coarser grained than the rimmed steel core treated similarly; of the two killed steels, the silicon-killed grade tends to be the coarser grained, although the difference is very slight.

ACKNOWLEDGMENTS

The authors wish to express their appreciation to Dr. E. C. Bain, who proposed this investigation, arranged for the materials used and contributed many valuable suggestions during the course of the work; to Mr. E. E. Legge and others of the American Steel and Wire Company, Worcester, who supplied the steels with the various reductions together with the chemical analysis and other data; to Mr. E. S. Davenport for guidance and helpful criticism; and to Messrs. C. E. Morgan, M. H. Pakkala and N. A. Matthews for their assistance with portions of the investigation.

DISCUSSION

Written Discussion: By John J. B. Rutherford, research metallurgist, The Babcock & Wilcox Tube Co., Beaver Falls, Pa.

Although the excellent manner of presentation is readily apparent on perusing this article, it is less easy to conceive the quantity of research necessary to acquire these data; it is hardly in order, therefore, to suggest that the authors might have continued their efforts to include, for example, silicon-killed steels of nominal phosphorus content, various types or degrees of aluminum-deoxidation, etc. One important feature of these steels has been overlooked

and that is the austenite grain size customarily related to 1700 degrees Fahr., as a reference point; the authors might rectify this omission in subsequent discussion.

Certain grades of low carbon steel, seamless tubular products are required to be manufactured to a hardness of Rockwell B 65 maximum. To make allowance for commercial variation, the fact that hardness measurements taken on the periphery are invariably higher than when taken on the cross section, and the slight hardening affected by straightening after annealing, possible aging, etc., it becomes evident that a "true" hardness of Rockwell B 55-60 must be obtained and this in a product free from exaggerated grain growth or pearlite patches. An aluminum-killed, fine-grained steel seems to be best suited to meet the above conditions as this permits more rapid coalescence of the carbide constituent on annealing after cold drawing. Silicon-killed steels, or what might be termed as partially deoxidized steels, which obtain an austenite grain size of No. 6 or coarser at 1700 degrees Fahr., have an "inherent softness" of about Rockwell B 65 which is very difficult to lower except by a long-time, high-temperature anneal which serves to decarburize the material. Stressing the higher creep properties obtainable in silicon-killed as contrasted to rimmed or aluminum-killed steels has resulted in current specifications requiring definite silicon additions which engenders difficulty in meeting hardness requirements where such steel is used.

The authors have shown that low carbon steels silicon or aluminum-killed, when cold-worked and then annealed at temperatures below A_1 for periods up to 16 hours, although little softening is accomplished after 2 hours, may be lowered to a minimum of Rockwell hardness B 60, except when influenced by exaggerated grain growth. This is interpreted to be the equivalent of Rockwell B 65 as measured on small tubular products. It would be of interest to learn of the experience of others concerning the hardness floor-level observed in these grades of steel. Incidentally, it is not believed that the Rockwell hardness requirement of B 65 for condenser and heat exchanger tubes is of any significance in application as the tubes exhibit great ductility at hardnesses of B 70 or 75, but the specification nevertheless is troublesome when silicon-killed steel is used.

Written Discussion: By G. L. vonPlanck, chief metallurgist, Columbia Steel Co., San Francisco.

The authors are to be commended for bringing to the light some of those intangibles which are associated with cold work and annealing. The phenomenon of grain growth upon annealing after critical strain is well known to mill operators, but hazy ideas exist as to what reduction percentages or annealing temperatures are necessary to avoid grain growth. We know that the data presented in this report will be of particular interest to those engaged in cold reduction of strip, bars, wire, etc. There are, however, other applications such as hot and cold rolling of sheets where similar problems are involved.

We personally would have appreciated more data covering reduction in area in smaller increments up to 30 per cent. In sheet mill production, "cold rolling" or "cold passing" after hot rolling is sometimes responsible for exaggerated grain growth although such cold reduction seldom exceeds 5 per cent. It is

usually impossible to trace back the actual percentage of cold work given the material prior to annealing, and we have the effect without knowing precisely the cause.

We believe it would have been better if the authors had picked samples of rimmed steel more closely approaching the analysis of the killed steel specimens for their investigation. The rimmed steel was of "low metalloid" type, and only represents to a limited degree the analyses of commercial rimmed steels. How much different the results would have been with higher carbon content is problematical. At least the extreme difference in recrystallization behavior between rim and core would have been minimized. Similar data on mechanically capped steels would also have been an interesting addition to the subject.

This report constitutes a splendid addition to the literature which brings together under one heading the many scattered data relating to the subject, and presents a clear-cut and concise description of the behavior of the various steels under similar conditions.

Written Discussion: By H. K. Work, Jones and Laughlin Steel Corp., Pittsburgh.

With the production of strip steel expanding so rapidly, the importance of the subject selected by the authors for their study is readily apparent. While considerable technical data are now available on the reaction of cold-worked low carbon steel to various annealing temperatures and annealing cycles, the authors made an important contribution to the general subject. The difference in behavior of rim and core to annealing treatment is particularly interesting, since it cannot be readily explained by mere difference in amounts of common elements between rim and core. The difference in behavior between aluminum-killed and silicon-killed steel under similar treatments is also significant and should stimulate further interest in this subject.

It is important to point out that the authors used cold drawing as a means of cold working the steels tested. In so far as this study may be applicable to wire processing no exception can be taken to cold drawing the test specimens. However, in applying conclusions drawn from these experiments to strip steel it is essential to bear in mind that extrapolation from cold drawing to cold rolling may not be entirely justified. It has been our experience that the effect of cold rolling is generally more drastic than that of cold drawing and exerts a more profound effect on the physical properties of the cold-worked material. Additional light may also be shed on this subject by extending the annealing cycle beyond the maximum range of 16 hours used in the present study, since in pot annealing of strip steel the annealing cycles are many times longer.

These remarks should not be taken as a criticism. The authors are probably fully aware of the limitations in scope of their present investigation, and we hope this paper is a forerunner of more extended studies on the same subject.

Authors' Reply

The authors are grateful to those who have discussed this paper. The suggestions offered as to other aspects of the general problem of

annealing cold-reduced steels that might be investigated are all excellent ones. We believe, however, that the necessity for setting certain limits to the investigation described in the paper is obvious and requires no apology. It is hoped that those interested in this problem will contribute additional information.

Mr. Rutherford raises the question of the austenitic grain size, as developed at 1700 degrees Fahr., of the steels investigated. Results of a carburizing test (8 hours at 1700 degrees Fahr.) are as follows:

Steel	Austenite Grain Size (ASTM)
Silicon-killed	3, few 5
Aluminum-killed	3, few 5
Rimmed—rim	3-4
Rimmed—core	5-6

The foregoing results show the two killed steels to be largely coarse-grained under the conditions of this test. Mr. Rutherford's comments as to the effect of degree of deoxidation of the steel on the rapidity of coalescence of the carbide constituent are very interesting.

Mr. vonPlanck mentions that the rimmed steel investigated is of the "low metalloid" type and is representative of only a small proportion of the rimmed steel produced commercially. This is true, and perhaps was not emphasized sufficiently in the paper. Caution should be used in applying the results obtained with this steel in processing other types of rimmed steel. We believe, however, that these other types of rimmed steel will show the same *qualitative* difference in behavior of rim and core on annealing as the low metalloid type investigated.

Mr. Work brings up the question of the relative severity of deformation produced by cold drawing as compared to cold rolling. We have no evidence of our own to offer on this point and appreciate the information contributed by Mr. Work.

HEAT ETCHING AS A MEANS OF REVEALING AUSTENITE GRAIN SIZE

BY M. J. DAY AND J. B. AUSTIN

Abstract

This paper describes a simple and rapid method for heat etching which has proved useful in revealing austenite grain size for a wide variety of steels over a wide range of heat treating conditions. The method depends upon preferential transfer of material away from grain boundaries, presumably by vaporization, when the metal is exposed to a high temperature in an inert atmosphere. The method has been successfully applied to a number of materials ranging from highly purified iron extremely low in carbon to high alloy steels; it does not appreciably alter the carbon content of the specimen, is rapid and requires only relatively simple apparatus. Direct comparison shows that it gives the same results as methods now in use. This is the only practical, noncarburizing method available today for delineating the austenite grain boundaries of steels with less than 0.1 per cent carbon. Grain growth in an aluminum-killed steel containing 0.07 per cent carbon has been investigated over the range 1700 to 2500 degrees Fahr.

INTRODUCTION

THE determination of austenite grain size involves in principle three distinct processes, establishing the grains under a given set of conditions, revealing their boundaries and rating their size. Of these, the process of rating is not influenced by any quality or behavior of the steel but is simply a matter of assigning to the grains as revealed some conventional number or symbol which characterizes their size. On the other hand, the result of the other two processes depends upon the behavior of the sample being examined and varies with a number of factors, notably composition of the metal and the heat treatment it has been given. This has led to the development of many different methods of establishing and revealing the austenite grains, some being adaptable to general use, others

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being designed for a specific set of conditions. These methods have been reviewed recently by Ward and Dorn,¹ whose summary shows that there is still a place for a method which can be applied to all grades of material from purified iron to the alloy steels, over a wide range of heat treating conditions without changing the chemical composition of the sample. One method, which is merely touched upon by Ward and Dorn, but which comes close to meeting the requirements stated, is preferential vaporization at the grain boundaries. This phenomenon was observed as early as 1909 by Rosenhain and Humfrey² and has been mentioned by numerous investigators³ since that time. The literature up to 1922 was reviewed by Hemingway and Ensminger⁴ who made further contributions of their own. They observed several networks on a polished surface of carbon steel which had been heated for six hours at 1000 degrees Cent. (1830 degrees Fahr.) in a vacuum and then furnace-cooled, and attributed these to the several sets of grains which had existed during the heating and cooling period. As this method revealed more than one system of grains and no other features except an occasional twin-band it was not considered satisfactory for grain size determination. Since 1922, the subject has been mentioned in several papers but as no careful evaluation of the method and its usefulness appears to have been made it was decided to explore its possibilities more fully. The results, which are described in this report, show that with proper technique the method is rapid, flexible, and widely applicable and that it requires only relatively simple apparatus.

APPARATUS

In order to obtain a suitable etch without changing the composition of the sample, it is necessary to have an atmosphere which is,

¹N. F. Ward and J. E. Dorn, "Grain Size of Steel," *Metals and Alloys*, Vol. 10, 1939, No. 3, p. 74.

²Rosenhain and Humfrey, *Proceedings*, Royal Society (London), 1909, A 83, p. 200.

³P. Oberhoffer, *Metallurgie*, Vol. 6, 1909, p. 542. Sauerwald, Schultze, and Jackwirth, *Stahl und Eisen*, Vol. 44, 1924, p. 1805. H. Hanemann, *Internationale Zeitschrift für Metallographie*, Vol. 3, 1912, p. 146. H. C. H. Carpenter and J. M. Robertson, "Structural Changes in Hypoeutectoid Steels on Heating," *Journal*, Iron and Steel Institute, Vol. 127, 1933, p. 259. A. Sauveur and C. H. Chou, "The Gamma-Alpha Transformation in Pure Iron," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 84, 1929, p. 350. Rawdon and Berglund, *Scientific Papers of the Bureau of Standards*, No. 571. G. Derge, A. R. Kommel and R. F. Mehl, "Some Factors Influencing Austenitic Grain Size in High Purity Steels," *TRANSACTIONS*, American Society for Metals, Vol. 26, 1938, p. 153. I. T. Baruzdin, *Metallurg* (Russian), Vol. 13, 1938, p. 13. Hugh O'Neill, "Alloy and Fine-Grained Steels for Locomotive Coupling Rods," *Journal*, Iron and Steel Institute (London), Vol. 135, 1937, p. 187.

⁴Hemingway and Ensminger, *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 67, 1922, p. 392.

among other things, neither oxidizing, deoxidizing, carburizing nor decarburizing. Moreover, it is desirable to have the period of heat treatment as short as is consistent with other requirements and to be able to quench the specimen to avoid any complication arising from slow cooling. A further advantage to be derived from quenching is that the surface becomes wrinkled, thus obliterating other networks which sometimes show as ghosts. These ghost networks are thought to be due to a slight etching of previously existing austenite grain boundaries which have left some trace of their presence upon the metal. They are never as deeply etched as the boundaries existing at the time of test, which may be etched until very deep furrows are made.

The requirements for a suitable inert atmosphere are reasonably met in practice by high vacuum, purified nitrogen, purified hydrogen, or a mixture of these two gases. High vacuum technique at heat treating temperatures is not simple and when the desirability of quenching in vacuum is considered as well, it becomes fairly difficult. For this reason, high vacuum is not considered as satisfactory as an inert gas. Nitrogen is satisfactory as far as ease of manipulation is concerned, and has no tendency to react with the sample, but it is not as easy to purify as hydrogen. Direct test shows, however, that with proper precautions it can be successfully used. Pure hydrogen is, of course, not in equilibrium with a steel containing oxygen or carbon but experiment shows that when carefully dried, hydrogen reacts at such a slow rate that it can be safely used as a substantially inert atmosphere, except perhaps at temperatures approaching the melting point. At these high temperatures oxygen diffuses rapidly through iron and treatment with hydrogen may deoxidize a layer deep enough to influence the grain size as observed on a polished surface. Experience with the hydrogen-reduction method of analysis for oxygen demonstrates, however, that except for an exposure of an hour or two at a temperature above about 2200 degrees Fahr. (1200 degrees Cent.) the loss of oxygen should not be appreciable, and, as such extreme conditions are the exception rather than the rule in determining austenite grain size, no difficulty from this cause should arise in most cases. Experience with the hydrogen-reduction method for oxygen has also demonstrated that there is an evolution of carbon monoxide at the higher temperatures. The effect of such a loss of carbon monoxide on the grain size is difficult to estimate but it is believed to

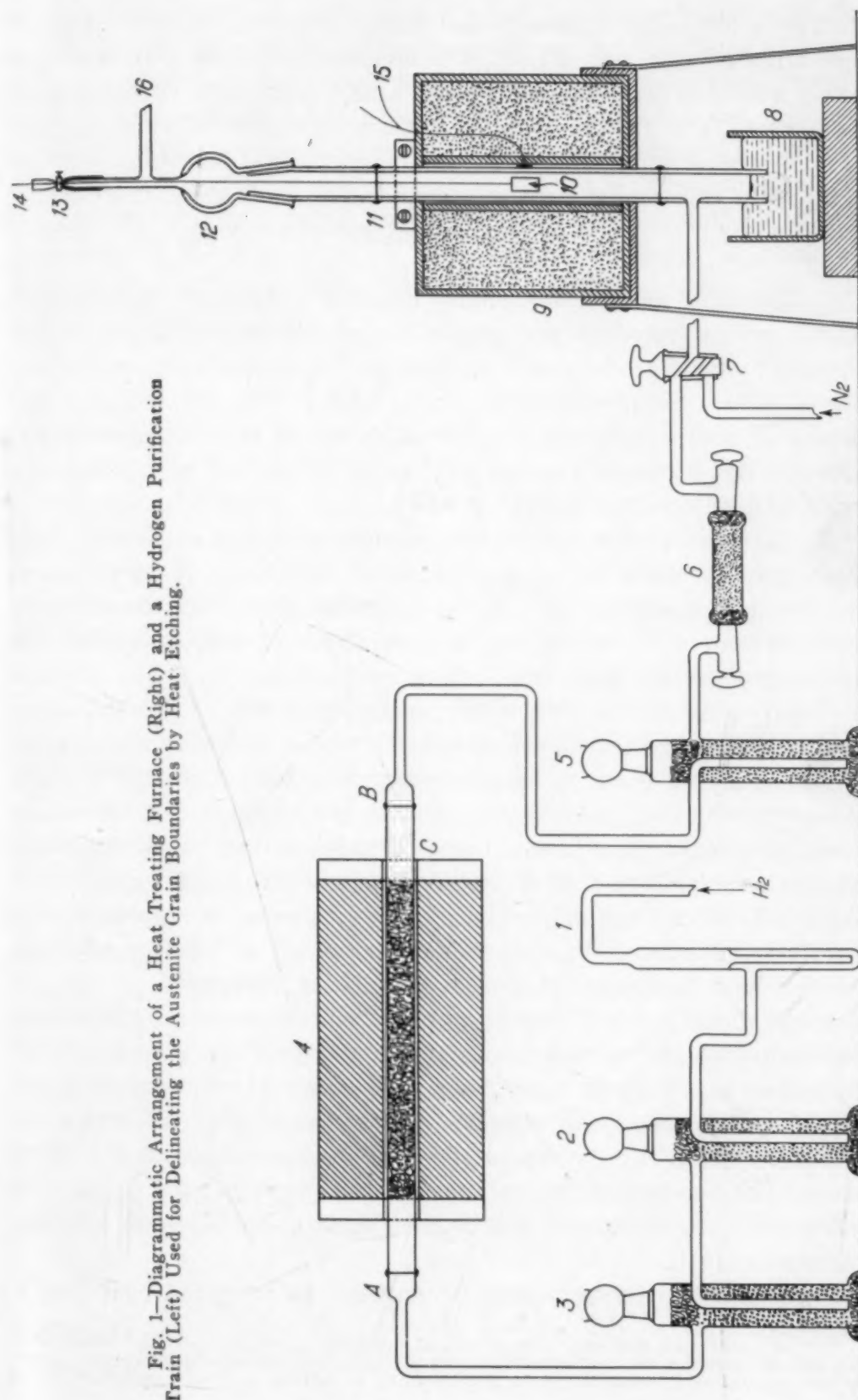


Fig. 1—Diagrammatic Arrangement of a Heat Treating Furnace (Right) and a Hydrogen Purification Train (Left) Used for Delineating the Austenite Grain Boundaries by Heat Etching.

be negligible under ordinary conditions. Mixtures of purified nitrogen and hydrogen are similar in behavior to hydrogen and have been used with success. They have the added advantage of being non-explosive. The gas obtained by cracking ammonia is, after purification, a convenient mixture of this kind. The determinations described in this paper were, however, made with an atmosphere of purified hydrogen which was found most convenient for general use.⁵

The most suitable quenching medium is mercury, which gives rapid cooling yet does not impair the metallographic polish of the specimen nor introduce any oxidizing or decarburizing vapor into the system. Simply dropping the specimen into the mercury and letting it float is sufficient in most cases, but if still faster cooling is desired, the specimen can be attached to the end of a rod or heavy wire and submerged in the mercury.

The apparatus as used with hydrogen is shown in Fig. 1. Tank hydrogen flows through a sulphuric acid bubbler (1), which serves as a rough flowmeter as well as a preliminary drier, then passes over Drierite (2) for further drying, and next over soda lime (3) to remove carbon dioxide. It then passes over platinized asbestos (4) to convert traces of free oxygen to water vapor. This catalyst, which is packed in a porcelain tube C, whose ends (A and B) are sealed to Pyrex glass, is heated to approximately 550 degrees Cent. After passing over the catalyst, the gas goes to a drying tower (5) containing activated alumina, thence through a tube containing phosphorus pentoxide (6) which reduces the residual water vapor to less than 0.00002 milligram per liter of gas at room temperature, and finally is introduced through a double stopcock (7) into the bottom of the heat treating chamber whence it is exhausted at the top through a capillary. The heating chamber is a porcelain tube placed inside a platinum-wound furnace, whose temperature is controlled by means of a thermocouple (15) as shown. The lower end of the porcelain tube dips into a mercury quenching bath; the upper end is fitted with a Pyrex cap, a piece of rubber tubing and a pinch clamp (13) to permit the specimen, which is suspended by an iron wire (14), to be lowered into the furnace or dropped into the quenching bath.

Satisfactory oxygen-free nitrogen can be prepared by bubbling

⁵Presumably tank hydrogen without special purification, or any controlled atmosphere, can be used provided that it does not produce a visible scale on the specimen when heated to the temperature in question for the time required to develop a visible network.

commercial nitrogen through a solution of 1-1 ammonium hydroxide saturated with ammonium chloride in contact with metallic copper, and then drying it over activated alumina and phosphorus pentoxide. A convenient and sensitive test for checking the removal of oxygen is to pass the nitrogen through a solution of cuprous chloride, which develops a blue color if oxygen is present. Another convenient test which is applicable to both nitrogen and hydrogen is to suspend a bright piece of 18-8 stainless alloy in the furnace. If oxygen or water vapor is present the surface soon becomes tarnished.

PROCEDURE

The specimen to be tested is given a metallographic polish and is suspended in the tube above the furnace. The air in the furnace is then flushed out with tank nitrogen, which in turn is purged with purified hydrogen, a procedure which considerably reduces the danger of an explosion. The sample is then lowered into the hot zone and held there for the desired heat treating period, at the end of which it is allowed to cool slowly or is quenched into mercury, as desired. With this arrangement a wide range of cooling rates, from the rapid quench in mercury to a furnace cool, is available. In general, however, a mercury quench is employed since, if the furnace is slow cooled, the sample is likely to show several networks whose interpretation may be uncertain. With the mercury quench on the other hand, micro-wrinkles appear on the polished surface and effectively wipe out all features except the deeply etched grain boundaries. Quenching the sample is therefore recommended.

During the run, the hydrogen in the furnace is kept at a pressure slightly above atmospheric to prevent air from entering through the exhaust capillary or the releasing clamp. Before removing the specimen from the furnace the hydrogen is flushed out with tank nitrogen. The test specimen as it is removed from the furnace is ready for rating the grain size. For certain other purposes, such as studying the products of transformation in relation to the austenite grains, it may be desirable to give the specimen a slight re-polish.

With the procedure described austenite grain size can be determined relatively quickly since the furnace can be kept hot at all times and, if the same temperature of test is used, specimens can

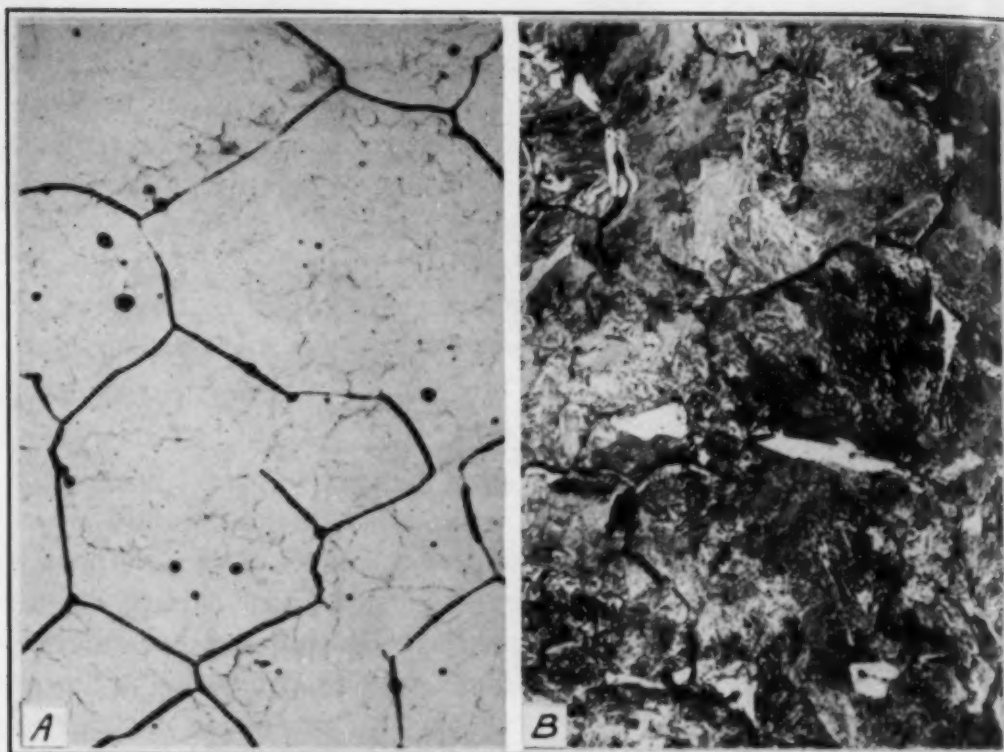


Fig. 2—Austenite Grain Size of a Eutectoid Steel Heated 24 Hours at 2000 Degrees Fahr. (1100 Degrees Cent.) and Furnace Cooled. $\times 250$.

A. As Heat Etched; Note the Clear Definition of the Austenite Grain Boundaries.
B. Same Specimen as (A) After Superimposing a Picral Etch. Observe the Slight Decarburization Produced Purposely to Show the Coincidence of the Austenite Grain Boundaries as Revealed by Heat Etching and as Delineated by Proeutectoid Ferrite.

be prepared at a rate determined only by the heat treating time desired. Moreover, the rate of production can be increased by suspending several specimens in the furnace at the same time, or by adding a suitable gas-lock at the top of the furnace to eliminate the flushing operation.

RESULTS

That the deeply etched boundaries obtained in this method are those of the austenite grains established at the high temperature is amply demonstrated by three facts: (1) the heat-etched boundaries coincide with the austenite grain boundaries as marked out by a ferrite network in hypoeutectoid steels and by the fine pearlite network of steels cooled at a rate slightly less than the critical cooling rate, (2) the agreement of the results with those of other non-carburizing methods, (3) the frequent occurrence of twinned grains; for while twinning has been observed in a few rare instances in alpha iron or ferrite, the frequent occurrence of twins is

characteristic of gamma iron or austenite. It is also clear that there has been no scaling or carburization (see, for example, Fig. 2). The possibility of removing oxygen, which has already been discussed, is so small as to be negligible in most cases, so that the chief question with respect to a change in composition is the extent of possible decarburization.

A check on decarburization was made by heating specimens of eutectoid steel in purified hydrogen for 24 hours at different temperatures up to 2000 degrees Fahr. (1095 degrees Cent.). These were furnace-cooled, slightly repolished and etched with picral. Only traces of ferrite were visible. The specimens were then sectioned and examined for depth of decarburization, which proved to be very small. It is, of course, possible to have a depletion of carbon at the surface without getting visible ferrite, and to this extent, a microscopic test is inconclusive in demonstrating that there was no loss of carbon. It can be stated, however, that the treatment in hydrogen does not produce appreciable decarburization as judged by metallographic methods. A typical example of the results of this test is given in Fig. 2. The photomicrograph Fig. 2A shows the grain boundaries as revealed by a 24-hour treatment at 2000 degrees Fahr. Fig. 2B shows a portion of the same surface with a picral etch superimposed on the heat etch.

The wide range of applicability of the method has been demonstrated by tests on purified iron extremely low in carbon, on carbon steels, a number of S.A.E. steels, and high alloy steels containing nickel, chromium, aluminum or manganese, or combinations of these elements. Typical results are shown in Fig. 3. Of particular interest is Fig. 3D which shows the grains in a steel containing chromium, molybdenum, nickel and vanadium, which are difficult to reveal by ordinary methods. It is clear that for such steels and for iron and low carbon steels this method has many advantages over those now in common use.

The reliability of the method in giving the same grain size rating as those now in use was checked by comparing the result of the heat etch on a steel containing 0.94 per cent carbon with that obtained by the gradient quench and grain-size etching methods. Two sets of runs were made, one with 11 minutes at temperature, the other with 60 minutes at temperature, each set including five temperatures between 1450 and 1850 degrees Fahr. A. S. T. M. grain size ratings were used throughout. Determinations by the

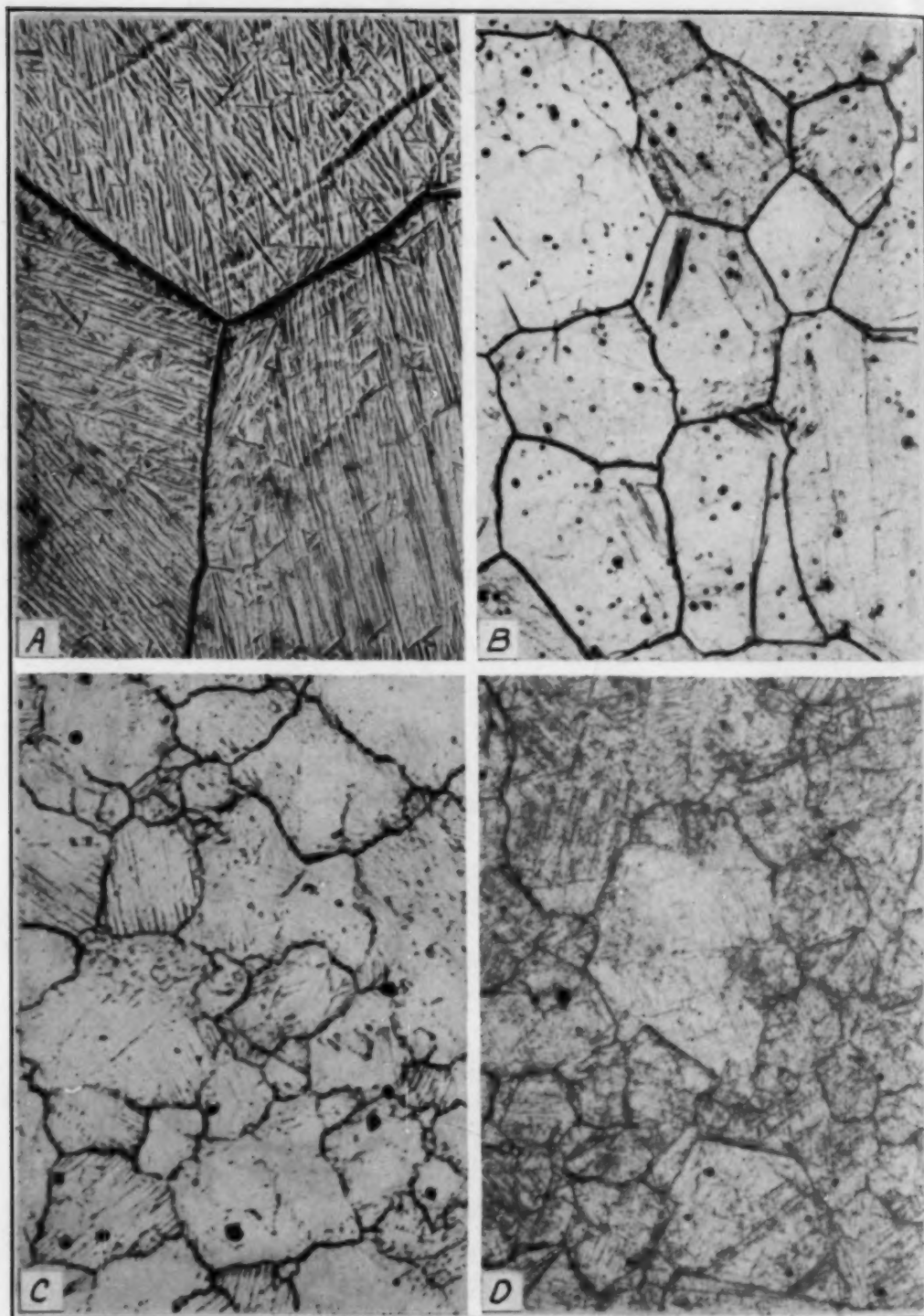


Fig. 3—Austenite Grain Boundaries of Highly Purified Iron, Plain Carbon Steels, and Alloy Steels as Revealed by Heat Etching. $\times 100$. Note Wrinkled Surface of the Specimens Quenched in Mercury and the Smooth Surface of that (B) Cooled in Hydrogen to Room Temperature. Twins Are Plainly Visible in (B).

- A. Hydrogen-Purified Iron, 1 Hour at 1700 Degrees Fahr., Quenched in Mercury.
- B. Hypoeutectoid Steel, 4 Minutes at 2000 Degrees Fahr., Cooled in Hydrogen.
- C. Hypereutectoid Steel, 1 Hour at 1900 Degrees Fahr., Quenched in Mercury.
- D. Alloy Steel Containing Cr, Mo, Ni, and V, 1 Hour at 2000 Degrees Fahr., Quenched in Mercury.

Table I
Comparison of Grain Size in 0.94 Carbon Steel as Revealed by Different Methods

Treatment	Grain Size (A.S.T.M. No.) at—				
	1450° F.	1550° F.	1650° F.	1750° F.	1850° F.
11 Minutes at Temperature					
Gradient quench	8-7	7	6	4	3
Heat etch	8-7	7	6-5	5	4-3
Tempering + grain size etch..	8-7	7	6	5-4	3
60 Minutes at Temperature					
Heat etch	8-7	7-6	5	4-3	3
Tempering + grain size etch..	8-7	7-6	6-5	4	3

gradient quench method were made on a separate set of specimens; determinations by the temper-and-etch method were made on the specimens used for the heat etch. After the heat-etched specimens had been rated the test pieces were tempered at 600 degrees Fahr., sectioned, repolished and etched with grain-size reagent (5 per cent HCl and 1 per cent picric acid in alcohol). The results, given in Table I, show that there is substantial agreement among the three methods. It should be noted that the grain size as determined within the sample by the temper-and-etch method is the same as that determined on the surface by the heat etching method.

Grain Growth in Low Carbon Steel—One of the several advantages of the method of heat etching is that it can be used as easily for low carbon steel as for any other material, which is not true of some of the more common methods. With such a steel, however, best results are obtained by quenching the specimen, since this treatment produces micro-wrinkles on the surface of each austenite grain which leave the boundaries clearly outlined, as is well illustrated in Fig. 4E.

As a demonstration of the usefulness of the method with low carbon steel, grain size was determined on aluminum-killed sheet containing 0.07 per cent carbon over a range of heat treatments

Table II
Austenitic Grain Size on Low Carbon Steel
 (A.S.T.M. Standard No.)

Temp. °F.	Time at Temperature in Minutes			
	10	30	90	1000
1700	9-8	8	8	6-5
1900	6	4	4-3	3-2-1
2100	3	2	1-2	1-0
2300	2	1-0	0	-1*
2500	1-0	-1*	-2*	

*Grain size numbers (-1) and (-2) are respectively equivalent to the commonly used (00) and (000) symbols.

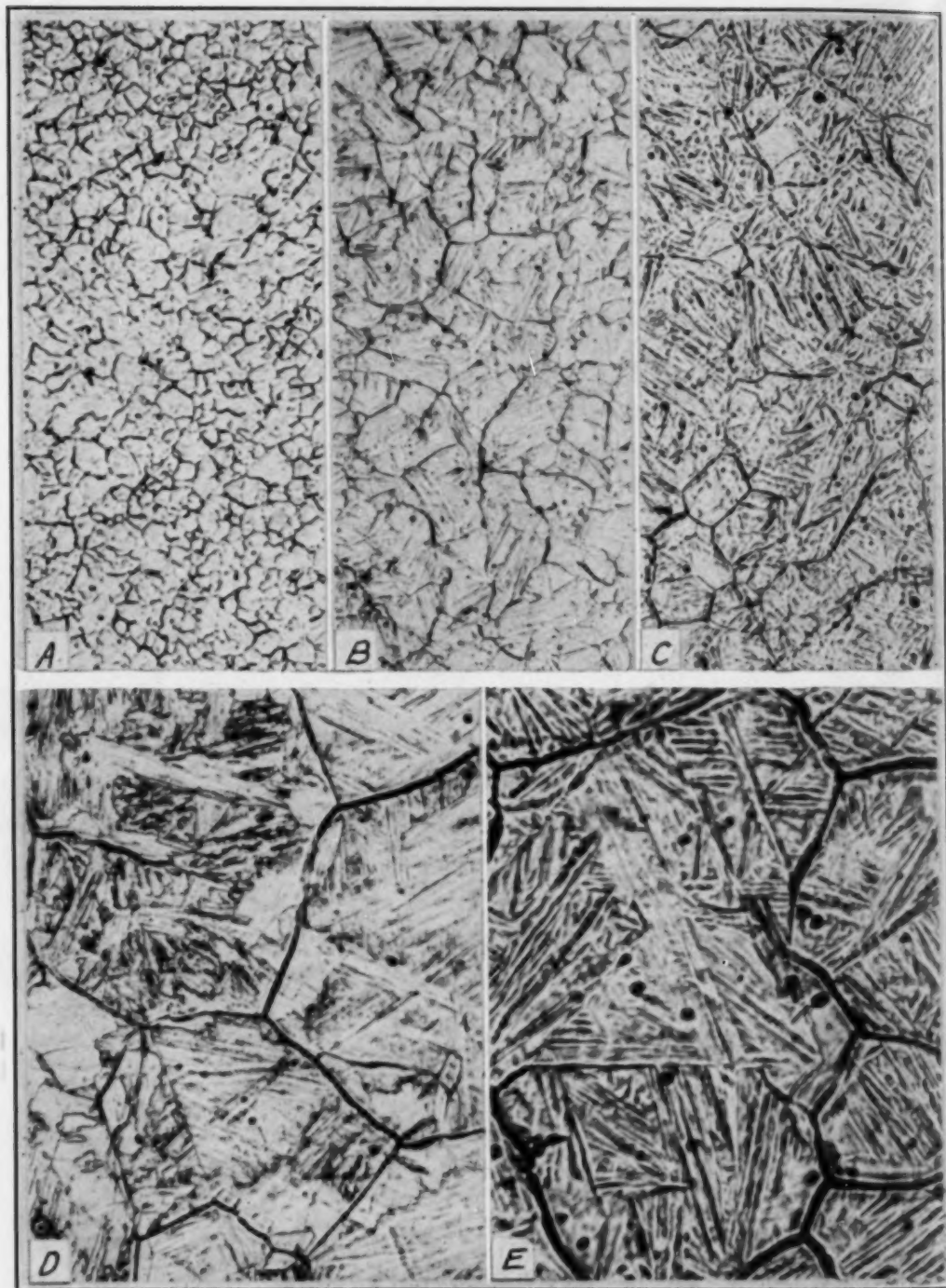


Fig. 4—Austenitic Grain Size of 0.07 Per Cent Carbon Steel Held for 30 Minutes at (A) 1700, (B) 1900, (C) 2100, (D) 2300, (E) 2500 Degrees Fahr. Quenched in Mercury. $\times 150$. Note Clear Definition of the Austenite Grains at all Temperatures.

from 10 to 1000 minutes at five temperatures between 1700 and 2500 degrees Fahr. The results are shown in Table II and in Figs. 4, 5, 6 and 7. Table II gives the A.S.T.M. grain numbers;

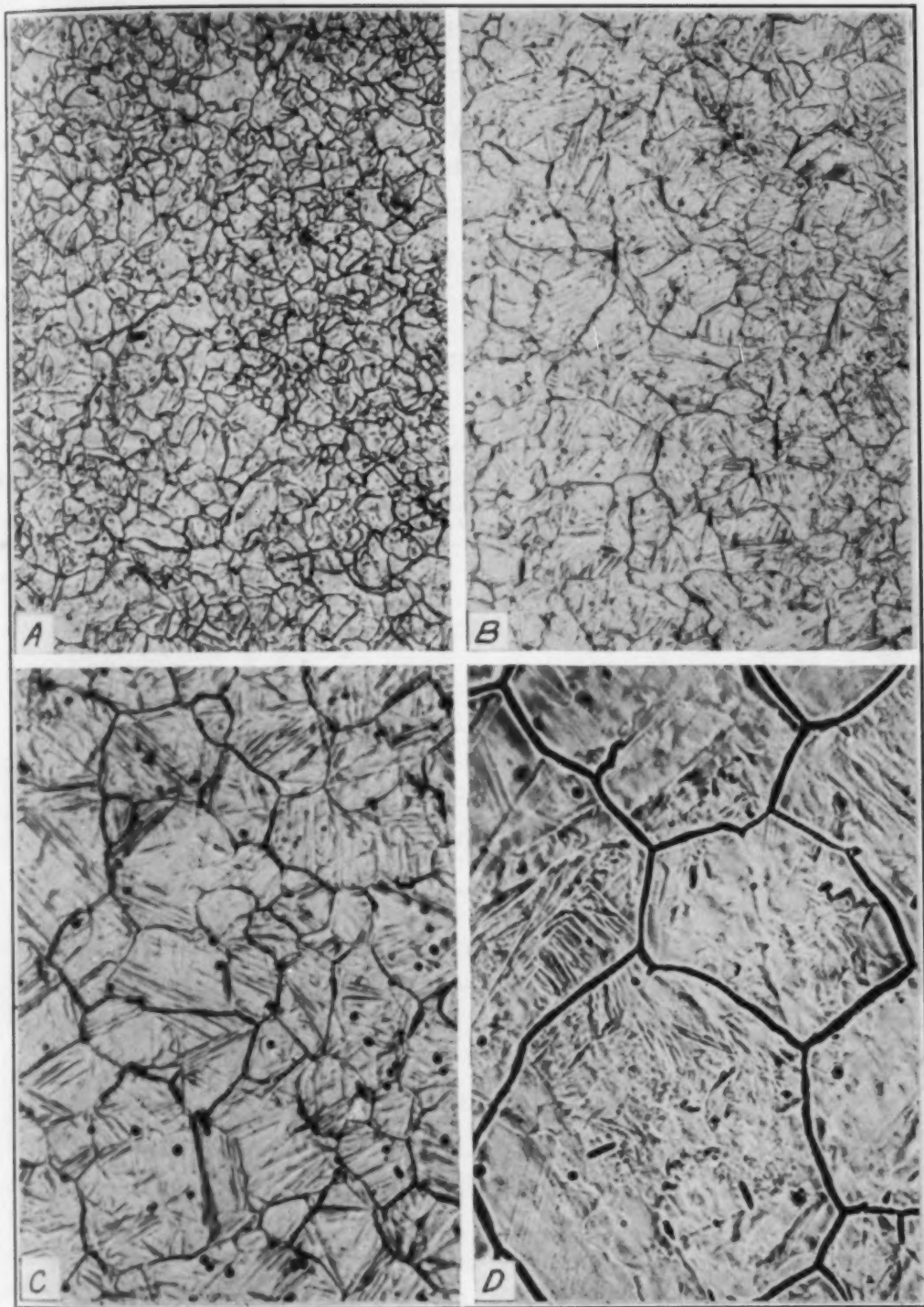


Fig. 5—Austenitic Grain Size of 0.07 Per Cent Carbon Steel Held at 1900 Degrees Fahr. for (A) 10 Minutes, (B) 30 Minutes, (C) 90 Minutes, (D) 1000 Minutes. Quenched in Mercury. All $\times 150$. Observe That the Austenite Grain Boundaries are Clearly Defined in as Short a Time as 10 Minutes.

Fig. 4 shows the appearance of the set of specimens held for 30 minutes at 1700, 1900, 2100, 2300, or 2500 degrees Fahr.; Fig. 5

gives a corresponding set of photomicrographs for samples held at 1900 degrees Fahr. for 10, 30, 90 or 1000 minutes respectively. The variation of mean area per grain with time at temperature of heat treatment is shown graphically in Fig. 6, which gives a good idea of the general behavior of this steel. The variation with temperature of mean area per grain for exposures of 10 and of 30 minutes

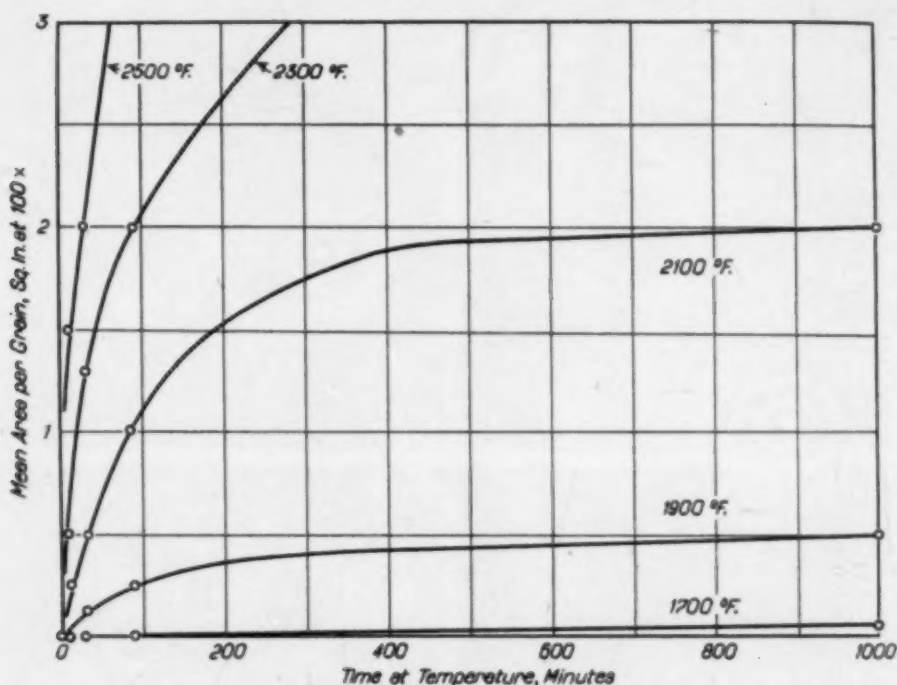


Fig. 6—The Effect of Time of Heat Treatment at Various Temperatures on the Austenite Grain Size of a 0.07 Per Cent Carbon Steel.

is shown in Fig. 7. This diagram is interesting because it shows that for this steel the term "coarsening temperature" has no precise meaning, unless it is qualified by some specification of time of heat treatment.

DISCUSSION

It may appear at first thought somewhat strange that a method which outlines the grain boundaries by transfer of metal should be able to follow grain growth so easily, since one would expect that the grain boundaries which are etched after exposure for 10 minutes should persist and be visible, at least as scars, after say 30 minutes exposure when the grains are much larger. This difficulty does enter occasionally but there are two factors which remove it in most cases. First scars or indications of the earlier grain boundaries are often visible in slowly cooled specimens. If, however, the specimen is quenched, the wrinkling of the surface obliterates all sur-

face structure but the boundaries existing at the time of quench; quenching is, therefore, equivalent to erasing the traces of former boundaries. Second, examination of a number of specimens indicates that although grain growth takes place in a few instances by a sort of radial expansion of a grain, it more often occurs as a coalescence of grains between which a boundary simply disappears; that is, grain growth is commonly not a process of uniform expan-

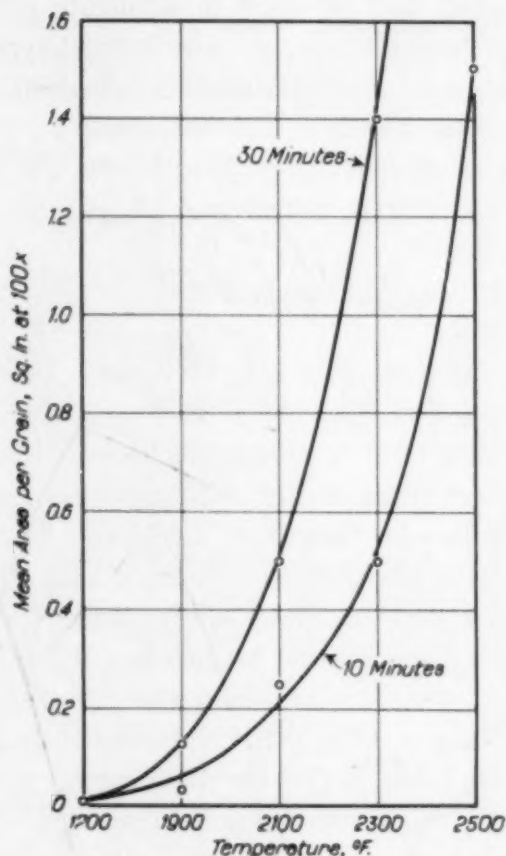


Fig. 7—Austenite Grain Size of a 0.07 Per Cent Carbon Steel Versus Temperature for 10-Minute and 30-Minute Heat Treatments. The Term "Coarsening Temperature" Applied to this Steel is Thus Meaningless Unless the Time of Heat Treatment is Specified.

sion of a grain, as is the case in the growth of pearlite colonies when austenite transforms below the eutectoid temperature, but is a process analogous to sintering. This behavior is illustrated in Fig. 2A, in which the disappearance of a boundary between two grains is clearly indicated in several places.

The wrinkling of the surface on quenching also deserves some comment. The wrinkles, which are obtained simply by rapid cooling without any etching, bear a strong resemblance to martensite

needles and are removed by a slight repolishing which does not influence the etched grain boundaries, which suggests that they are due to the presence of a thin skin of martensite or some similar product over the surface. A further indication that this is the case is the fact that the wrinkles are more pronounced in carbon steels and are not so marked in highly purified iron.

Although most useful in determining austenite grain size in steel, the method can also be used to delineate grain boundaries in any metal at any temperature at which this type of transfer of material, presumably by preferential vaporization, takes place at a reasonably rapid rate. It can, for instance, be used to determine ferrite grain size at temperatures just below the A_1 , and is applicable likewise to nonferrous metals and alloys.

CONCLUSIONS

Heat etching which causes preferential removal of material from the grain boundaries has been studied as a method of revealing austenite grain size. A simple technique has been developed which has been used successfully with purified iron, hypoeutectoid steels, eutectoid steels, hypereutectoid steels and high alloy steels over a wide range of heat treating conditions. The method consists in holding a polished specimen in an inert atmosphere at the desired temperature for the desired time and then quenching it in mercury, which procedure marks the austenite boundaries at the time of quench unambiguously. This method is the only practical noncarburizing method available today for delineating the austenite grain boundaries of steels with less than 0.1 per cent carbon; it yields identical results with other noncarburizing methods where comparison is possible. The decarburizing action of carefully purified hydrogen was found to be so slight as to leave the carbon content of the specimen substantially unaltered. The usefulness of the method with low carbon steel is demonstrated by a study of grain growth in an aluminum-killed steel containing 0.07 per cent carbon over the temperature range 1700 to 2500 degrees Fahr. The results show that for this steel the term "coarsening temperature" is not precise unless the time at temperature is specified.

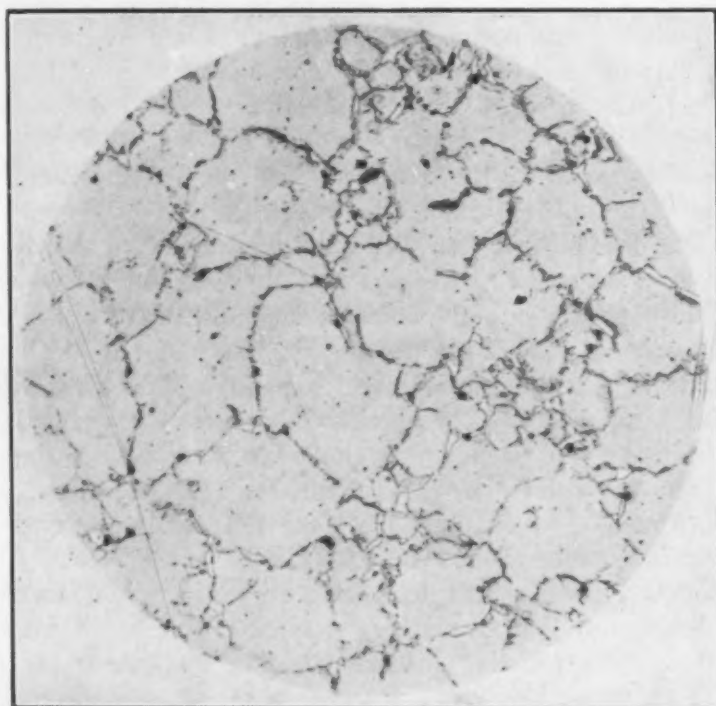
It is a pleasure to acknowledge our indebtedness to Mr. J. R. Vilella for his continued interest and advice and to Mr. H. A. Brown for help in preparing specimens.

DISCUSSION

Written Discussion: By N. F. Ward, associate professor, and J. E. Dorn, assistant professor, Department of Mechanical Engineering, University of California, Berkeley.

The investigations of Day and Austin on heat etching as a universal method of determining austenitic grain size introduces a new tool for grain size studies on steels which do not respond adequately to methods currently available. The method of quenching in mercury to obliterate all ghost furrows due to previous positions of the austenite grain boundaries permits an unambiguous delineation of grain size, previously impossible by preferential vaporization at the grain boundary without this added technique.

Although the heat etching method is essentially a laboratory technique that may be seldom employed in plant testing, it presents a method of investigating



Micrograph of an S.A.E. 1035 Steel Showing Austenite Grain Size by Gradient Quench.

the effect of grain size in a field where our information is quite inadequate. We are yet unaware as to the practical importance of grain size in low carbon steels that are not carburized and in high alloy steels of high hardenability. Most of our present information on grain size refers to carburizing grades of steel, plain carbon steels, and low alloy steels that show pronounced differences in mechanical properties upon hardening.

In their conclusion there appears to be an implication that we believe the authors did not intend to incorporate in their manuscript, namely that grain growth "occurs as a coalescence of the grains between which a boundary

simply disappears." Since the grain boundary demarks a transition region wherein the atoms are no longer located on regular lattice sites but exhibit a gradual transition from the sites of the lattice of one grain to those of the adjacent grain, a disappearance of the boundary is tantamount to a complete rotation of one grain, en masse, until both grains have the same lattice orientation. This is statistically improbable.

Incomplete grain markings are frequently encountered in the gradient quench method but are seldom found in the carburizing or slow cooling methods. We believe that this indicates that incomplete markings are due to the fact that some boundaries are less active than others and consequently remain undelineated in cases where rapid cooling technique is employed. This difference in activity may be due to segregation.

Note occasional incomplete boundary in photomicrograph which may be due to differences in reaction rates along the boundary. This interpretation is emphasized by the rather large quantities of ferrite at some points where reaction rates were greatest, by fine pearlite at points where the rates were intermediate, and by undelineated boundaries where the reaction rates were a minimum. It is suggested that segregation may be the cause of these differences.

Written Discussion: By Hugh O'Neill, chief metallurgist, London Midland & Scottish Railway Co., Derby, England.

In 1939 Ward and Dorn published a critical review of the methods available for revealing austenitic grain size, but they overlooked a procedure devised by the present writer and published during 1937 in a paper mentioned by the present authors. In this paper the simplicity of the McQuaid-Ehn test was recognized, but attention was drawn to possible difficulties associated with the carburizing compound and the previous history of the steel. It was therefore suggested that high temperature etching could be used, the possibilities being "(a) vacuum etching, (b) gas etching, and (c) thermoelectronic emission. The last method would permit of the use of the electron microscope but the apparatus would need a certain amount of development. Alternatively, X-ray spectrograms from a high-temperature camera would enable the grain sizes to be determined." Eventually a vacuum-chlorine hot etching method was perfected using a gradient furnace enabling the grain size at different temperatures for a given time to be determined on the one specimen. The results obtained by vacuum and vacuum-chlorine etch were compared with those from the McQuaid-Ehn, the quench and fracture and the quench and etch methods, and the present writer therefore questioned the authors' general statement that "no careful evaluation of the method and its usefulness appears to have been made."

The hot hydrogen etch and quench test which the authors had devised appeared to be a very satisfactory method for studying austenite grain size, and it could be utilized in a gradient furnace is desired. At the time of his own experiments the writer had deliberately avoided a hydrogen atmosphere partly because of curious twinning effects noted by Rawdon and Berglund (Bureau of Standards, 1928), but more particularly because of decarburization and grain size changes described by Austin (*Journal, Iron and Steel In-*

stitute, London, 1922, I). The evidence given in Table I of the present paper now indicated, however, that hydrogen had not caused any spurious results, and the authors were to be congratulated on an interesting and valuable piece of work.

Oral Discussion

O. W. McMULLAN:⁶ There is one question I would like to ask the authors in regard to the slowly cooled samples which show a slight veining within the larger network. Do you consider that to be the pearlite grain structure or, I should say, the alpha iron grain structure?

Authors' Reply

We are grateful to Messrs. Ward and Dorn for their discussion and particularly for calling attention to a possible misinterpretation of our remark on coalescence of grains because it gives us an opportunity to remove any misunderstanding of our views on the subject. We were referring to the observation that occasionally a grain boundary is either not delineated or has been obliterated for one reason or another and we do not wish to suggest a mechanism of grain growth about which, indeed, we hold no definite views.

It is also a pleasure to have Mr. O'Neill's comments, coming as they do from a pioneer in this field. Our statement regarding a careful survey of the subject was not intended as a criticism of his, or any others' work, but was an expression of the conviction which we had in starting this investigation that a critical re-examination of the method would lead to useful results. In connection with his remarks on possible decarburization we should like to add a word or two about the treatment of several specimens of different carbon content at the same time. If the hydrogen used is carefully dried there is little likelihood of appreciable transfer of carbon from one specimen to another unless the time of treatment is far longer than any we have used. On the other hand if a trace of moisture is present so that some reaction between gas and metal may occur fairly rapidly, a transfer of carbon may result. In general, we believe it inadvisable to treat samples of widely different carbon content at the same time unless precautions are taken to insure that the atmosphere is genuinely inert.

In reply to Mr. McMullan we should not like to commit ourselves definitely on the nature of the veining within the larger network, but we incline to the belief that it is associated with transformation products formed during slow cooling.

Since the presentation of the paper we have found one or two specimens which did not coarsen even in a temperature range within which grain growth was confidently expected. A re-examination of such specimens beginning with a new polish and including the most careful control of experimental conditions has invariably led to satisfactory results, which indicates that certain samples may be sensitive to small changes in procedure. In consequence, we advise anyone who may find an apparent failure of the method to repeat the treatment under more careful control or with a slightly different polishing technique.

⁶Youngstown Sheet & Tube Co., E. Chicago, Ind.

THE EFFECT OF HUMIDITY OF AIR ON THE OXIDATION OF A LOW CARBON STEEL

BY C. A. SIEBERT AND H. G. DONNELLY

Abstract

This paper deals with the effect of water vapor in air on the oxidation of a low carbon steel in the temperature range 1825 to 2150 degrees Fahr. (995 to 1175 degrees Cent.). It has been found that steels oxidize to a greater extent in humidified air than in dry air. The ferrous iron content of the scale formed at any given temperature in this range is higher when humidified air is used than when dry air is used, which results in a faster rate of diffusion and therefore a greater scaling loss.

A SERIES of papers (1), (2), (3), (4), (5)¹ have been published in the TRANSACTIONS of the American Society for Metals dealing with the work carried out in our laboratories on the oxidation of steel. The present paper deals with the effect of the humidity of the air on the scaling of a low carbon steel in the temperature range 1825 to 2150 degrees Fahr. (995 to 1175 degrees Cent.). The literature on the subject of oxidation has been reviewed in the publications mentioned above; therefore no review will be given at this time.

The experimental work was carried out in tubular, resistance-wound electric furnace, the tube being approximately 2 inches in diameter and 30 inches long. A closed-end thermocouple protection tube and an air inlet tube were sealed into one end of the furnace, the thermocouple protection tube being so placed that the hot junction of the thermocouple was directly over the sample during a run. The closed end of the furnace tube was packed with broken porcelain for several inches to aid in raising the temperature of the air to the furnace temperature. Tests were run under two conditions at each temperature, namely:

1. Using dried and purified air for the oxidizing medium.

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, C. A. Siebert is assistant professor of metallurgical engineering, University of Michigan, Ann Arbor, Mich., and H. G. Donnelly is instructor in chemical engineering, Wayne University, Detroit. Manuscript received June 16, 1939.

2. Using purified and humidified air for the oxidizing medium.

Drying and purification of the air was accomplished by passing it through towers of calcium chloride, ascarite, and dehydrite. Humidification, for condition No. 2, was accomplished by mixing air and low pressure steam, cooling the mixture to 68 degrees Fahr. (20 degrees Cent.) and removing the entrained moisture in a series of traps. Tests showed that the air was saturated with water vapor at this temperature of 68 degrees Fahr. i.e., the air contained 0.0147 pound of water vapor per pound of dry air, or 1.45 per cent by weight. The flow of air was so regulated that a linear velocity of 30 feet per minute was maintained through the furnace at all temperatures, assuming that the air reached the temperature of the furnace.

The samples were cylinders 2 inches long and 0.625 inch in diameter, and were machined from $\frac{3}{4}$ -inch stock and polished through 00 emery paper to the final dimension of 0.625 inch. The analysis of the steel was as follows:

C	Mn	P	S	Si	Cr	Ni
0.14	0.50	0.016	0.022	0.25	0.07	0.14

The samples suspended on a chromel wire cradle were introduced into the furnace through which the stream of air was passing, and the total time in the furnace was 1 hour. In general, 5 to 7 minutes were required for the specimen to come to temperature. At the conclusion of a run, the samples were immediately quenched in water to eliminate further scaling on cooling as well as to prevent any change in scale analysis due to the eutectoid transformation in the iron-oxygen system.

The scale analysis was determined by dissolving a sample of scale in dilute sulphuric acid under an atmosphere of carbon dioxide, titrating the ferrous iron with standard 0.1 normal KMnO_4 , reducing the iron with aluminum coils, and titrating again with KMnO_4 to obtain the total iron content.

RESULTS

The scope of the investigation was limited to the effect of humidity of air on the oxidation of an S.A.E. 1015 steel. The data obtained for both conditions, i.e., scaling in dry air, and in humidified air saturated at 68 degrees Fahr., are summarized in Fig. 1.

The curves of Fig. 1 show that at all of the temperatures investigated, namely from 1825 to 2150 degrees Fahr. (995 to 1175 degrees Cent.), the humidified air produces a greater scaling loss than the dry air. The loss in weight when scaling in dry air at 1825 degrees Fahr. (995 degrees Cent.) is 0.085 pound per 100 square inches, and the oxidation loss increases with increasing temperature to a maximum value of 0.175 pound per 100 square inches at 1975

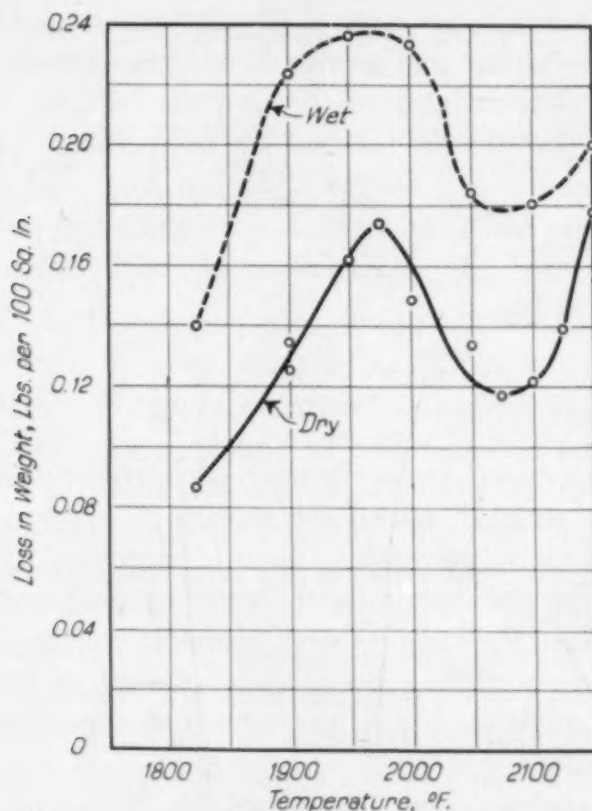


Fig. 1—Ferrous Iron in Scale Versus Temperature.

degrees Fahr. (1080 degrees Cent.). Between the temperatures of 1975 and 2075 degrees Fahr. the loss in weight decreases with increasing temperature, attaining a minimum value of 0.117 pound per 100 square inches at 2075 degrees Fahr. (1135 degrees Cent.). At temperatures above 2075 degrees Fahr., oxidation again increases with increasing temperature. When using humidified air saturated with water vapor at 68 degrees Fahr. the loss in weight at 1825 degrees Fahr. (995 degrees Cent.) is 0.14 pound per square inch, and the scaling loss increases with increasing temperature to a maximum value of 0.238 pound at 1975 degrees Fahr. (1080 degrees

Cent.). Between the temperature of 1975 and 2075 degrees Fahr. the loss in weight decreases with increasing temperature to a minimum value of 0.179 pound at 2075 degrees Fahr. (1135 degrees Cent.). Above 2075 degrees Fahr. the oxidation loss again increases with increasing temperature.

The variations in ferrous iron content of the scale formed at various temperatures is shown in Fig. 2. The temperatures at which the maxima and minima in the scale analysis curve occur correspond

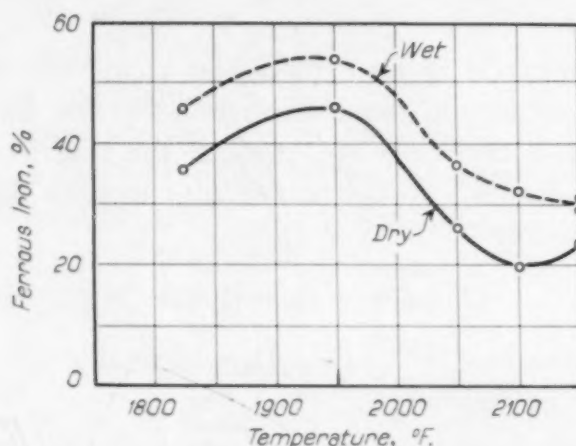


Fig. 2—Scaling Versus Temperature.

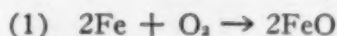
rather closely with temperatures of the maxima and minima of the loss in weight curves (Fig. 1). The ferrous iron content of the scale formed at any given temperature is greater when humidified air is used in place of dry air as the oxidizing medium; and likewise, the loss in weight at any given temperature is greater when humidified air is used in place of dry air as the oxidizing medium.

The phenomenon of the amount of scaling actually decreasing with increasing temperature was reported on by Upthegrove and Murphy (6), Siebert and Upthegrove (7), (8), and Siebert (9). As pointed out in previous publications, this decrease in scaling with increasing temperature is due to a difference in the degree of oxidation of the scale itself. Figs. 1 and 2 of this paper, as well as the previous work cited, show that as the temperature at which maximum scaling occurs is exceeded, a decreasing proportion of ferrous iron in the scale is observed. This decrease in ferrous iron content changes the character of the membrane through which diffusion of oxygen, in some form, must take place. It also changes the solubility of oxygen in the scale, as FeO has a wider solubility range than

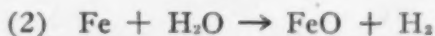
Fe_3O_4 which alters the driving force (concentration gradient) in the diffusional process.

It is evident from the above that at any given temperature, any change in the scaling conditions, which will increase the proportion of ferrous iron in the scale, will result in an increase in the scaling loss. At all of the temperatures investigated, humidification of the air resulted in an increase in ferrous iron content of the scale compared with that produced when dry air was used as the oxidizing medium. Also the scaling loss was greater in humidified air than in dry air.

In order to arrive at an explanation as to why the presence of water vapor in air should cause an increase in the ferrous iron content of scale, it is necessary to consider the reactions taking place when oxidation occurs. In the presence of oxygen the first reaction occurring is probably



and the FeO then oxidized to Fe_3O_4 and finally to Fe_2O_3 . Water vapor will react as follows:



and the FeO is then oxidized to a higher degree of oxidation. When a film of oxide has once been formed on the surface of the steel, the oxidizing gas must be dissolved at the surface of the scale in order to allow diffusion to take place. Water vapor as well as oxygen must dissolve in the scale and diffuse inward toward the metal-scale interface, while iron diffuses outward from the metal-scale interface. The double diffusion of iron and oxidizing substances results in setting up a reaction zone within the scale, and reaction (2) with the liberation of H_2 will tend to keep the scale in a lower degree of oxidation. The work of Upthegrove and Murphy (10) showed that the ferrous iron content of scale formed at temperatures ranging from 1800 to 2000 degrees Fahr. (983 to 1095 degrees Cent.) in an atmosphere of water vapor is approximately 59 per cent, while scaling in dry air in this same temperature range resulted in ferrous iron contents ranging from 55 per cent at 1800 degrees Fahr. (983 degrees Cent.) and falling rapidly to 21 per cent at 2000 degrees Fahr. (1095 degrees Cent.). The results obtained by Upthegrove and Murphy cannot be compared directly with the present investigation since their oxidizing atmosphere was 100 per cent water vapor;

however it does show that water vapor produces scale of high ferrous iron content.

SUMMARY AND CONCLUSIONS

The results of this investigation have shown that:

1. Scaling in humidified air produces a greater oxidation loss than scaling in dry air at all temperatures investigated, namely 1825 to 2150 degrees Fahr. (995 to 1175 degrees Cent.).
2. The scale formed in humidified air has a higher ferrous iron content than scale formed in dry air at any given temperature within the range investigated.
3. The higher scaling loss in humidified air compared to the loss in dry air is due to the higher ferrous iron content of the scale produced in humidified air.
4. The fallacy of conducting scaling experiments without controlling the humidity is very evident.

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6. Upthegrove and Murphy. Ibid.
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DISCUSSION

Written Discussion: By R. H. Harrington, research metallurgist, General Electric Co., Schenectady, N. Y.

This is one of those rare research problems which may be dealt with concisely, with clarity and an unusual degree of finality, so that, when finished, one may feel, "Well, that is that." The authors are certainly to be congratulated for their presentation.

The writer would make only one very small issue and that is within the variable meaning of phraseology and, therefore, of no great importance. The expression "The double diffusion of iron and *oxidizing substances*" would imply *molecular* diffusion in a crystalline solid if the "oxidizing substance" is other than atomic oxygen, itself. As an example: It is rather well recognized now that the compound Fe_3C does not molecularly diffuse through the iron lattice while the element carbon does atomically diffuse through the iron lattice. Another example is the atomic diffusion of hydrogen through iron and steel.

Written Discussion: By L. C. Whitney, chief metallurgist, Copperweld Steel Co., Glassport, Pa.

We have carried out experiments to study the oxidation of steel using an almost identical set-up to that described by the authors. Our work has been confined to the one temperature of 1800 degrees Fahr. but using various heating periods up to eight hours. Experiments were conducted both with dry and with humidified air and, as would be expected, greater amounts of oxide were developed in both cases with the longer heating periods. An interesting point which we observed, however, was that the difference between the amounts of scale developed in dry and wet air was the same regardless of whether the heating period was one hour or eight hours. For heating periods above five hours there was a tapering off in the rate of scale formation with the longer heating periods.

Our experiments were carried out on a steel of very similar analysis with the one exception that the manganese content was increased to about 2 per cent. It was interesting to note that with this increase of manganese content in the steel, the results which we obtained were in quite good agreement with those which the authors show in Fig. 1. This was particularly true with humidified air on which we obtained a value of 0.136 pound per 100 square inches of surface at 1800 degrees Fahr. as compared to 0.14 pound per 100 square inches of surface at 1825 degrees Fahr. as shown by the authors. With dry air, we obtained a value of 0.052 pound per 100 square inches of surface as compared to 0.085 pound per 100 square inches of surface reported by the authors. These data indicate very definitely that increasing the manganese content of the steel up to 2 per cent did not affect the rate of oxidation in humidified air at 1800 degrees Fahr. but it does appear to have reduced the amount of scale formed at this same temperature in dry air.

In our experiments, we have occasionally encountered samples on which the scale blistered and this invariably had a pronounced effect on the results obtained. The condition was more prevalent in dry air than in the humidified air. I should like to ask the authors whether they have encountered this same condition in their experiments and, if so, what was done to overcome it?

Authors' Reply

We wish to thank the people who have discussed our paper. In reply to Dr. Harrington we would like to say that the reason we used the words "oxidizing substance" was to eliminate the confusion of whether the oxygen had come from the air or from the water vapor, and we appreciate the fact

that undoubtedly it is oxygen that is diffusing and not water vapor, and if we left that impression it was an oversight on our part.

Mr. Whitney said that he found very good agreement in the wet air but not in the dry air with our particular results. We would say that in itself it was surprising because we have run various low carbon steels at times in our past investigations and find that practically every heat of steel would give us a different type of curve. It is possible that all of the variations in results we obtained on different steels may more or less have come together if we had used humidified air. All our previous information was obtained on dry air.

We have also noticed blistering to quite some extent. Usually the blistering appears about at the temperature where the maximum in the curve occurs. At the higher temperatures where the loss in weight curve is decreasing toward the minimum, the amount of blistering seems to increase. We feel from our studies that the blisters are regions where the FeO content of the scale is higher than the surrounding regions. We tried to isolate sections to get analysis by chipping off pieces of scale in the vicinity of a blister, but this does not give very accurate results.

In one of our previous publications we showed that the decrease in oxidation to a minimum was accompanied by this blistering effect, and that we felt it was definitely tied up with the decrease in the ferrous iron content, which alters the rate of diffusion of oxygen in the scale. Mr. Whitney points out that blistering is more prevalent in dry air than in humidified air. Our scale analysis shows that the FeO content of scale formed in humidified air is greater than the FeO content of scale formed in dry air at any given temperature. This tends to support the above explanation for blister formation.

THE MECHANICAL PROPERTIES OF THE ISOTHERMAL DECOMPOSITION PRODUCTS OF AUSTENITE

BY M. GENSAMER, E. B. PEARSALL AND G. V. SMITH

Abstract

A plain carbon eutectoid steel was reacted isothermally at temperatures ranging from the critical to 375 degrees Cent. (705 degrees Fahr.). The interlamellar spacing of the pearlite was measured down to 500 degrees Cent. (930 degrees Fahr.). At 475 degrees both pearlite and bainite were present; at lower temperatures no pearlite was observed. Measurements were made for each reaction temperature of complete tensile properties, fatigue strength, and cohesive strength. The results are plotted as functions of the reaction temperature and the pearlite spacing.

INTRODUCTION

IT would be of value to know the mechanical properties of each of the constituents observed in alloys, as a function of the size and mode of distribution of the particles of the phases. This paper reports the first results in a program of systematic study on this subject. The properties of mixed microstructures can be understood only when the properties of the individual and separate microstructures are known. It was decided to begin with eutectoid steel. The mechanical properties of steels given commercial heat treatments represent the properties of mixed microstructures; for during ordinary commercial heat treatment, austenite transforms over a range of temperatures, with each temperature contributing its own characteristic microstructure. Some scattered results of tests on the properties of steel with controlled and uniform microstructures have been reported (1),¹ (2), (3), (4), but the tests have been confined to hardness and the ordinary tensile properties. It would be desirable to know the whole stress-strain curve, well into the plastic range, for each structure. It would also be desirable to

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, M. Gensamer is associate professor of metallurgy and a staff member of the Metals Research Laboratory, Carnegie Institute of Technology, Pittsburgh, and E. B. Pearsall and G. V. Smith are research assistants in the Metals Research Laboratory. Manuscript received June 26, 1939.

know something about the resistance to the propagation of a crack as revealed by fatigue tests and measurements of cohesive strength. These three properties, namely, (a) stress-strain curve, (b) fatigue strength, and (c) cohesive strength, adequately describe the mechanical behavior of metals. For a complete description, the influence of plastic deformation on the fatigue strength and cohesive strength should be reported.

It was hoped that it would be possible to measure these three properties for a series of specimens consisting entirely of pearlite of controlled and uniform interlamellar spacing. Such specimens might be prepared by the strictly isothermal decomposition of austenite (5), and this method was attempted. It is suspected that the specimens prepared were not strictly uniform as to pearlite spacing. This may be caused either by failure to maintain a constant temperature during the reaction (for fast reactions the heat of the reaction may cause appreciable recalescence) (6), or by the circumstance that the pearlite spacing is only statistically a function of the reaction temperature. The authors have been unable to decide what range of interlamellar spacings exists in their specimens, and so where the spacing is referred to it is not intended to imply that it is strictly uniform. That the range in interlamellar distance is not great may be concluded from the agreement in the measured values, as determined by the two methods used. One of the methods simply takes the spacing as the minimum observed in the specimen, and the other measures an average spacing by a simple statistical method. This latter method involves a determination of the fraction of the surface that has a spacing greater than a certain value; the method is now being studied in this laboratory and will be the subject of a paper to appear in the near future, so it will not be discussed further here. Until this method has been verified, the spacings reported in this paper are to be considered as tentative.

MATERIAL AND HEAT TREATMENTS

The steel was of the grade used for bridge wire. It was found to be slightly hypoeutectoid; its analysis follows: 0.78 per cent carbon, 0.63 per cent manganese, 0.014 per cent phosphorus, 0.030 per cent sulphur, 0.18 per cent silicon. In the range of quenching temperatures used, it had an austenite grain size of 8 to 9, A.S.T.M. classification, as measured on the decarburized edge of the slowly

cooled specimens. Specimens were heated in a lead bath to the quenching temperature, which averaged about 825 degrees Cent. (1515 degrees Fahr.). A study of the effect of the quenching temperature revealed no significant effect within the range used, from 790 to 840 degrees Cent. (1455 to 1545 degrees Fahr.), although the higher temperature seemed to tend to lower the cohesive strength. The time at the quenching temperature was 0.5 hour. The specimens were quenched individually into another lead bath at a temperature known hereinafter as the reaction temperature. This lead bath was controlled to within ± 1 degree Cent., except for the short time interval immediately following the introduction of the hot specimen. In some instances the temperature rose as much as 5 degrees above the control temperature; within 5 minutes the temperature returned to within 1 degree of the control temperature. The time at the reaction temperature varied from 1 minute at 540 degrees Cent. to 20 hours at 706 degrees Cent. (1305 degrees Fahr.) and 45 minutes at 370 degrees Cent. (700 degrees Fahr.). For the longer times several specimens were in the bath at the same time. The specimens were finally cooled either in powdered insulating refractory to room temperature, requiring about 1 hour to reach a temperature below 100 degrees Cent., or in the lead bath to room temperature. After the latter treatment the lead bath was melted, reaching a temperature of 325 degrees Cent., then the specimens were removed and finally cooled in air. These two treatments gave no different results in any of the properties measured.

It is realized, of course, that the temperature of the specimen during the reaction may be quite different from that of the lead bath, especially when the reaction is fast. This difference should be a function of the size of the specimen. Normally the specimens had a diameter at the time of heat treatment of 0.260 inch. Several specimens of diameter 0.125-inch were transformed at several temperatures and tested for tensile strength. Within the limits of accuracy of these experiments, the small specimens had the same tensile strength as the larger ones. This, coupled with the fact that several of the test results when plotted against the temperature of the reaction fall on a straight line, as in Fig. 5, whether the reaction be fast as at 550 degrees Cent. or slow as at 650 degrees Cent. or 400 degrees Cent., is evidence that the reaction temperature is in no instance much different from that of the lead bath as reported.

Typical structures are illustrated in Fig. 1. Photomicrographs

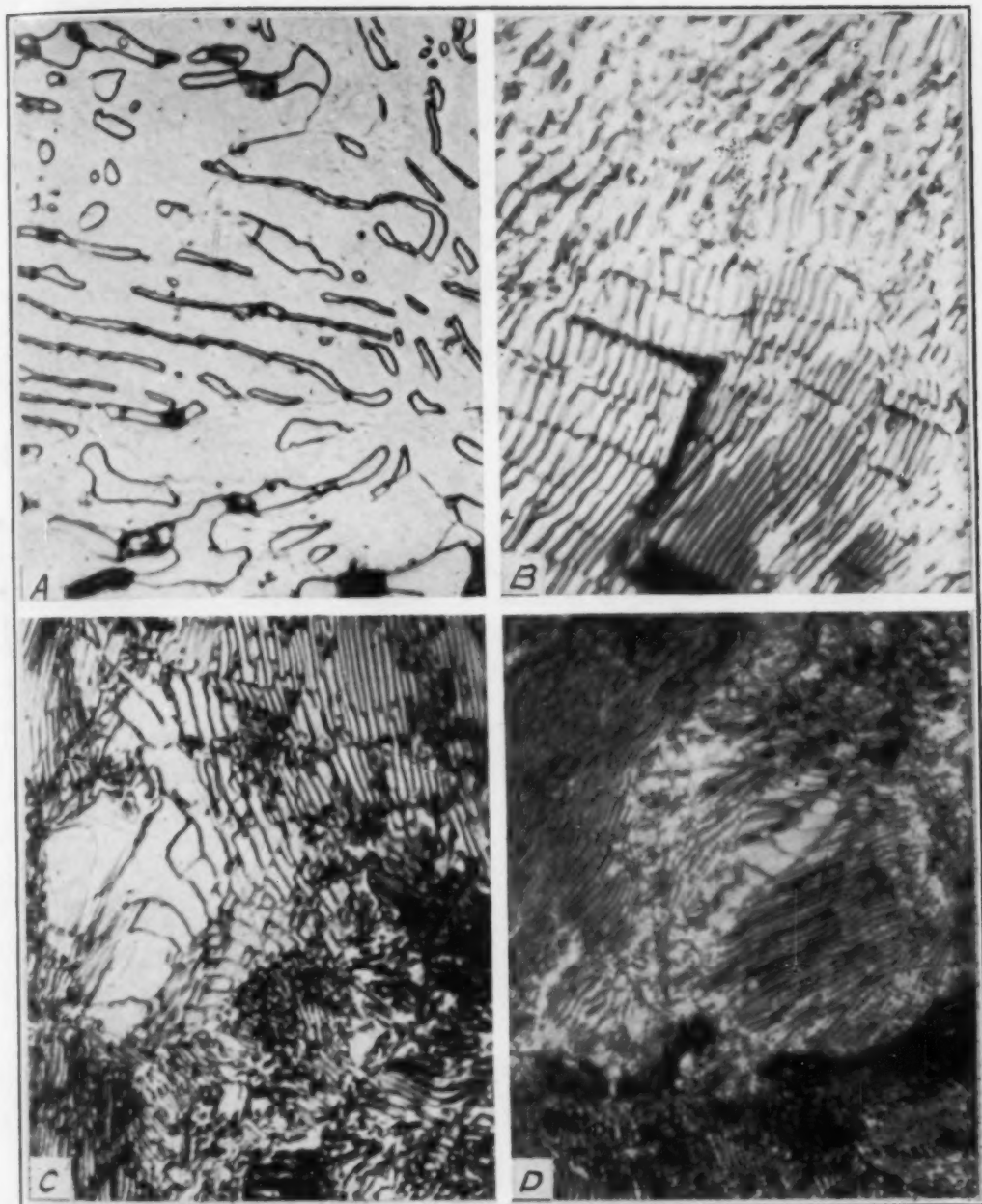


Fig. 1—Typical Microstructures. Reaction Temperatures as Follows: A—706 Degrees Cent.; B—684 Degrees Cent.; C—662 Degrees Cent.; D—620 Degrees Cent. $\times 2000$.

1A and 1B were taken near a tensile fracture, which accounts for the cracks. The interlamellar spacings obtained are plotted in Fig. 2 against the reaction temperature. The crosses are the values indicated by the least spacing observed; the circles are the values obtained by the application of the method using the relative areas above and below a certain spacing. In Fig. 2, the logarithm of the spacing, x , is plotted against the inverse of the absolute temperature,

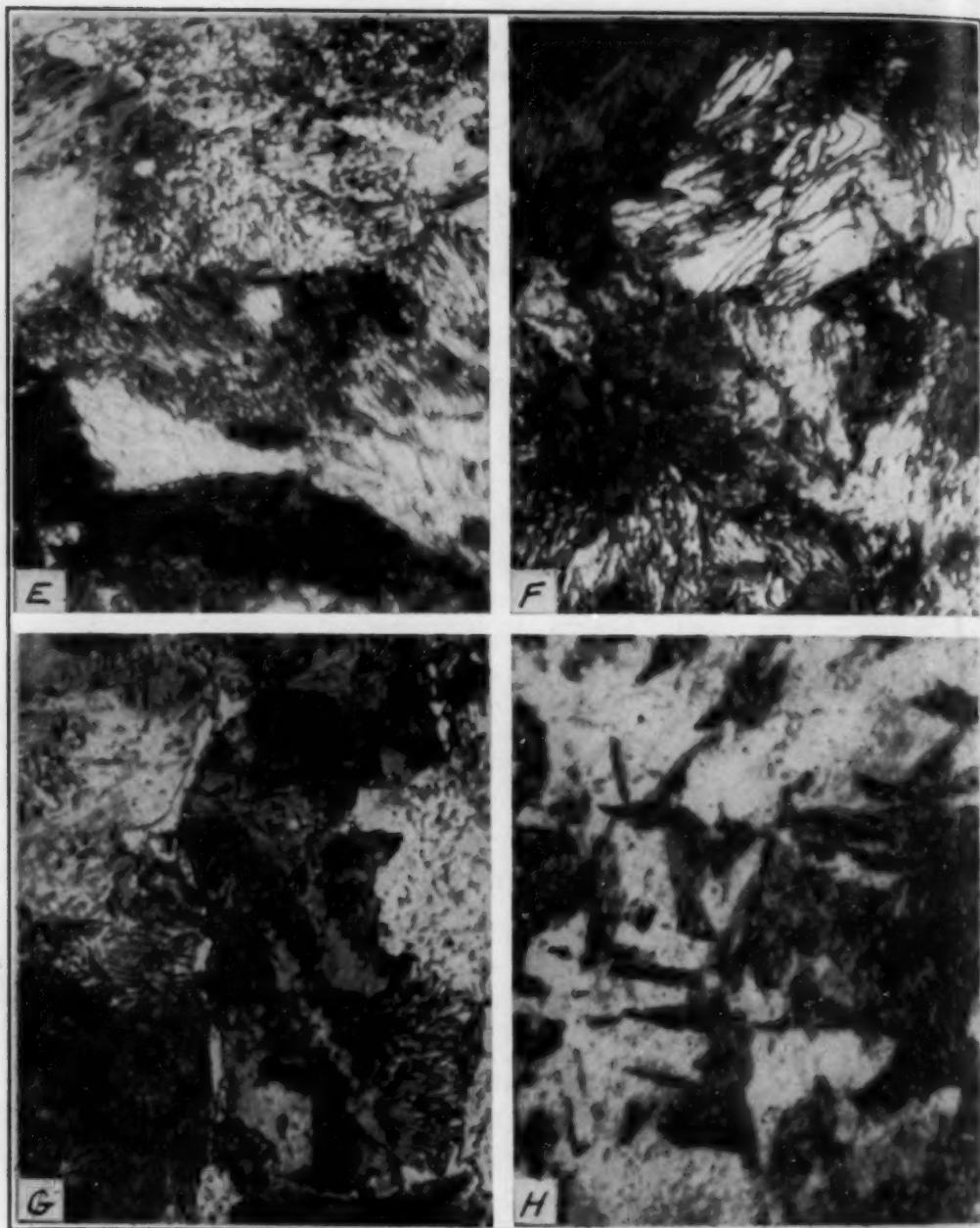


Fig. 1A—Typical Microstructures. Reaction Temperatures as Follows: E—583 Degrees Cent.; F—501 Degrees Cent.; G—475 Degrees Cent.; H—370 Degrees Cent. $\times 2000$.

T, as suggested by R. F. Mehl (Ref. loc. cit.). A straight line is also obtained by plotting $\log x$ directly against T (not its inverse). Over such a narrow temperature range as the one used, it is not possible to decide which is the correct plot. The $1/T$ plot was chosen because there is considerable theoretical justification for such a relationship.

TESTING TECHNIQUE

The tensile specimens when heat treated were 3 inches long; the fatigue specimens, 2 inches long. Each specimen had a threaded extension to fit it into a holder designed to hold it down in the bath.

The tensile specimens were machined to have a gage length of 1 inch and a diameter of 0.260 inch, with shoulders $\frac{5}{16}$ inch in diameter. After treatment they were threaded and the gage length machined to a diameter of 0.250 inch. This eliminated any effect of decarburization. The specimens were mounted in modified Robert-

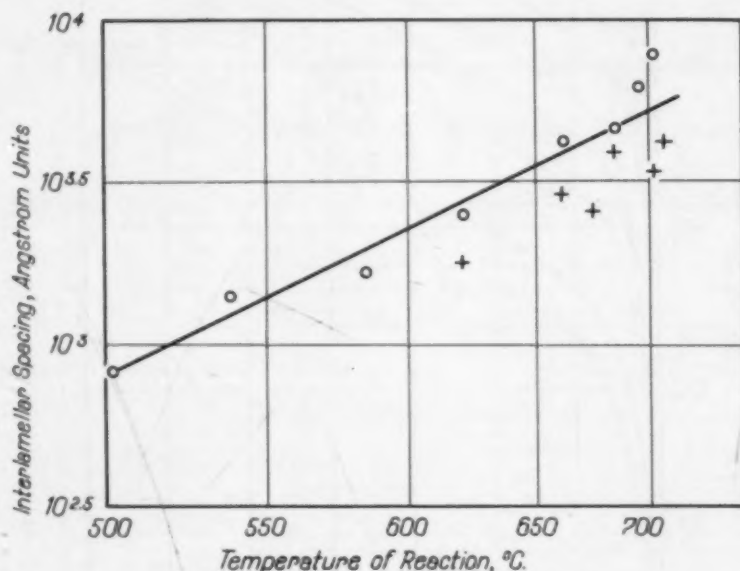


Fig. 2—The Relationship between Pearlite Spacing and Reaction Temperature. The logarithm of the spacing in Angstrom units (millimeters $\times 10^{-7}$) is plotted against the reciprocal of the absolute temperature. In a similar plot with the temperature instead of its reciprocal, the points fit a straight line almost as well as above. Crosses represent measurements made taking the minimum spacing observed, the circles measurements made of average spacing by a statistical method.

son shackles and adjusted to axial loading, using two Tuckerman optical strain gages of 1-inch gage length. Data for a stress-strain curve were obtained, using the optical strain gages to an extension of almost 1 per cent. Beyond this, the least diameter was measured with a pointed micrometer caliper, continuing to fracture. The rate of straining in the elastic range was about 10^{-4} inches per inch per minute. Approximately the same rate of cross head motion was maintained to fracture. The "true" stress, σ , was determined by dividing the load by the corresponding least actual area. The effective deformation ϵ was calculated from the change in diameter

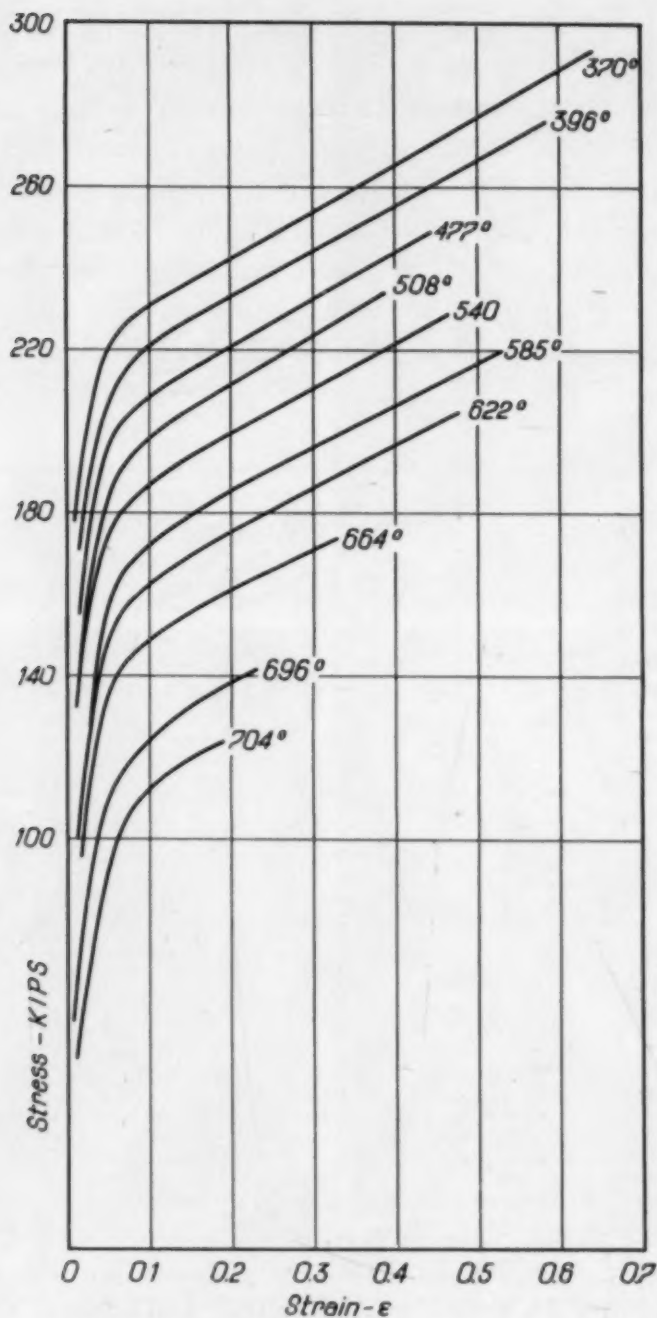


Fig. 3—"True" Stress-Strain Curves for Various Reaction Temperatures. The stress, σ , in thousands of pounds per square inch, is the load divided by the least actual area; the effective deformation ϵ , is given by the expression $\epsilon = \int \frac{dl}{l} = 2 \ln \frac{d_0}{d}$, where d is the least diameter.

by the formula $\epsilon = 2 \ln \frac{d_0}{d}$. (7) Some of these σ vs. ϵ curves are reproduced in Fig. 3.

The fatigue specimens were machined to a diameter of 0.260 inch, heat treated, then machined to a minimum diameter of 0.150 inch (some of the specimens run at the highest stresses were 0.140 inch), with a radius of 1.5 inch. They were polished through 000 metallographic polishing paper, finishing with longitudinal scratches. The machine used was a 10,000 revolutions per minute rotating cantilever beam type. The S-N curves obtained are plotted in Fig. 4.

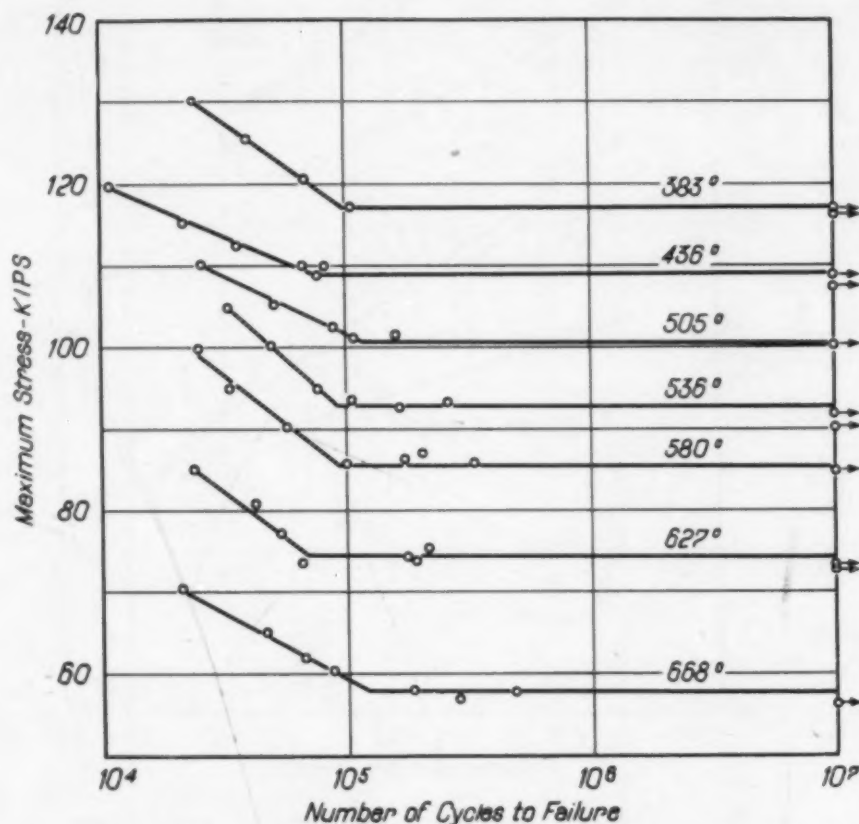


Fig. 4—Fatigue S-N Curves for Various Reaction Temperatures. The fatigue strength, in thousands of pounds per square inch, was obtained with a 10,000 r.p.m. cantilever beam machine of 40 inch-pounds capacity.

The cohesive strength specimens before treatment were identical with the fatigue specimens. After treatment they were machined to a diameter of 0.25 inch, threaded for a distance of 0.25 inch from both ends, and provided with a sharp 60-degree notch around the specimen at the center. The diameter at the base of the notch was $\frac{5}{32}$ inch. They were rotated under load in the fatigue machine until a crack was produced, completely encircling the specimen at the base of the notch. The machine was equipped with a microswitch so arranged that the additional deflection obtained when a crack

formed stopped the machine. The cracks were also observed with a magnifying glass, focused on the tension side of the stationary specimen under a slight load. These cracked specimens were broken in tension, using a fixture designed to prevent bending of the specimen.

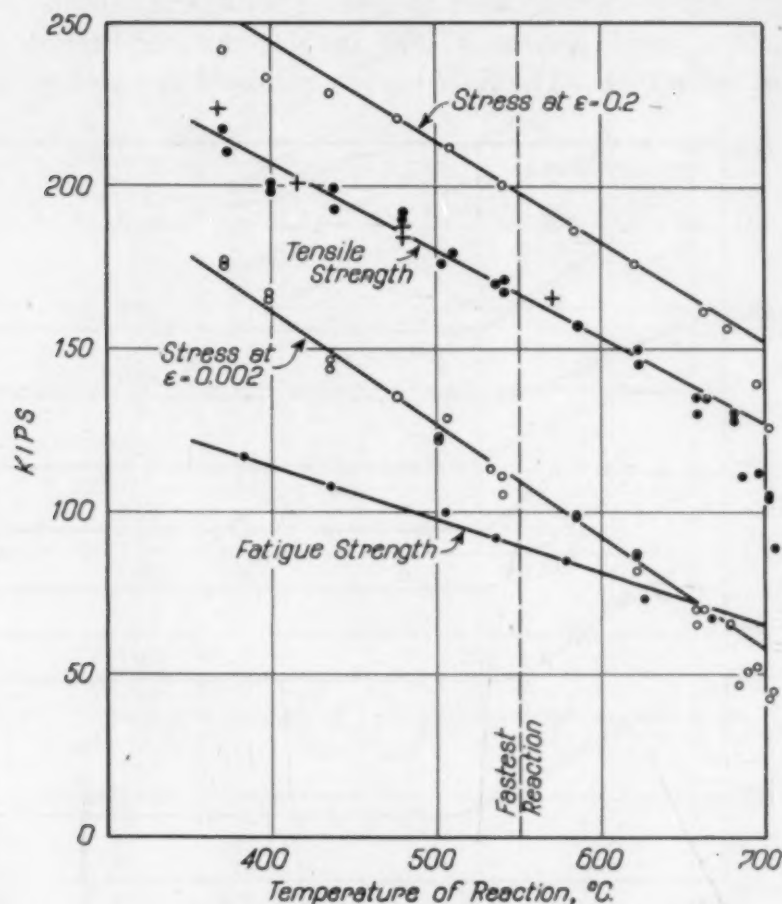


Fig. 5—Tensile Strength, Yield Strength, and Fatigue Strength for Various Reaction Temperatures. Strength in thousands of pounds per square inch; temperature in degrees Cent. The curve marked "stress at $\epsilon = 0.002$ " is the ordinary 0.2 per cent permanent extension yield strength. The curve marked "stress at $\epsilon = 0.2$ " is the true stress for which the total effective deformation ($\epsilon = 2 \ln \frac{d_0}{d}$) is 0.2; for all temperatures, this is on the straight line part of the curves of Fig. 3.

The specimens broke with little if any distortion, as indicated by a perfectly regular rate of increase in load up to a sudden fracture. The cohesive strength (perhaps more properly called, after Kuntze, the "technical cohesive strength") (8) was calculated by dividing the load at fracture by the area of the specimen remaining after the fatigue crack had been produced. The area of the tension crack was measured on a photographic plate using a planimeter. The appear-

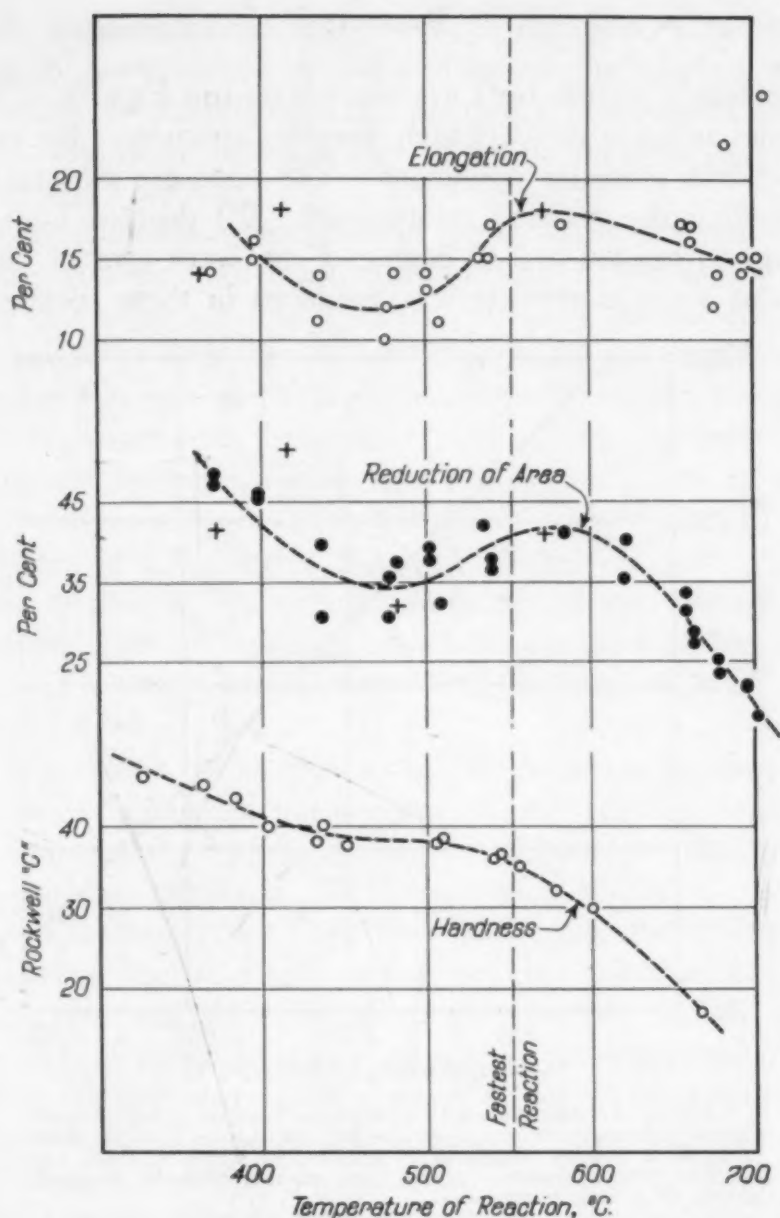


Fig. 6—Elongation, Reduction of Area, and Hardness for Various Reaction Temperatures. Elongation and reduction of area are in per cent at fracture. Elongation is measured over a 1-inch gage length on a specimen 0.25 inch in diameter.

ance of the crack produced in the fatigue machine is quite different from the appearance of the crack produced in tension. The fatigue cracks were occasionally not concentric with the axis of the specimen; these specimens usually gave lower values of the cohesive strength and were rejected. This technique and the variables influencing the values obtained will be the subject of a paper to be published in the near future.

RESULTS

The results of the tests are summarized in Figs. 5, 6, 7 and 8. The circles refer to the 0.25-inch diameter specimens, the crosses to the 0.125-inch diameter specimens. The tests are plotted down to and including the reaction temperature, 370 degrees Cent. Tests on specimens reacted at 330 degrees Cent. were erratic. It is suspected that some martensite was produced in these specimens.

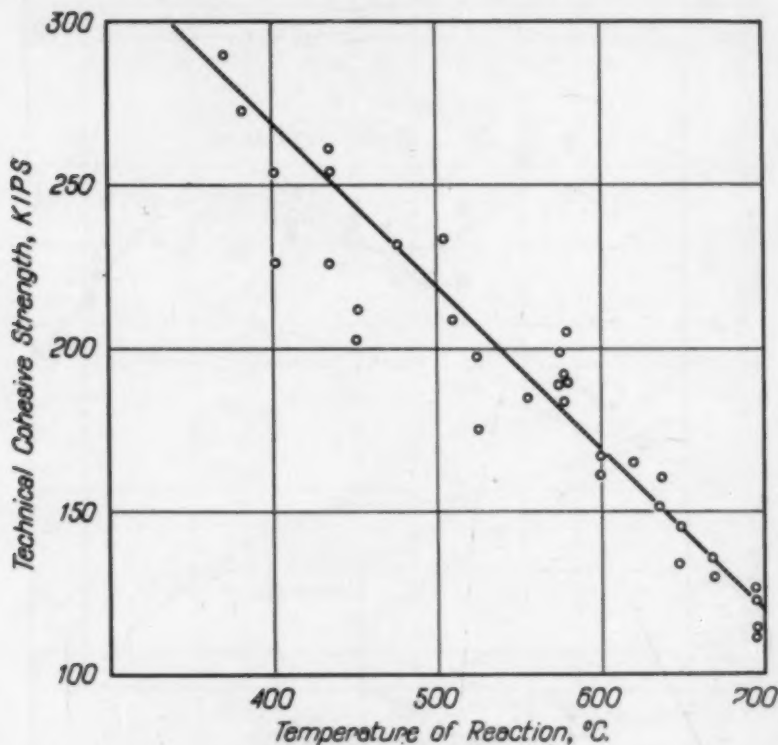


Fig. 7—Cohesive Strength at Various Reaction Temperatures. Cohesive strength is stress in pounds per square inch required to break a notched specimen in tension, in which a crack has previously been produced at the base of the notch by repeated loading in a fatigue machine.

It will be observed that in spite of the fact that the inter-lamellar spacing of the pearlite is not a linear function of the reaction temperature (Fig. 2) the plots against temperature of tensile strength (Fig. 5), yield strength (stress at $\epsilon = 0.2$ and $\epsilon = 0.002$ in Fig. 5), fatigue strength (Fig. 5), and cohesive strength (Fig. 7) are straight lines that extend even to low reaction temperatures for which at least part and probably most of the reaction product is the acicular structure which has come to be referred to as bainite. The fatigue strength is approximately one half the tensile strength only for the higher temperature reaction products; the ratio is

higher for the finer structures. Rockwell hardness versus temperature (Fig. 6) is definitely curved and agrees qualitatively with the work of the U. S. Steel Research Laboratory (Davenport and Bain, loc. cit.). The dashed line in Figs. 5 and 6 indicates the temperature of fastest reaction.

The reduction of area curves (Fig. 6) show an interesting lack of ductility for the high reaction temperatures, verified by the observation that these specimens tend to break at surface marks such as gage marks or scratches. This is probably associated with the low cohesive strength of the high temperature reaction product.

The departures of the plotted points from the curves at the higher reaction temperatures are undoubtedly caused by spheroidization, which it has been found impossible to avoid. Holding a specimen 4 hours at 820 degrees Cent. (1510 degrees Fahr.) prior to quenching produced less spheroidal cementite on reaction at 677 degrees Cent. than the usual $\frac{1}{2}$ -hour treatment at 820 degrees Cent. and yielded tensile test results more nearly in line with lower reaction temperature results.

It is interesting to note that the rate of strain hardening, after necking-down begins, is independent of the reaction temperature. This is indicated by the parallelism of the straight sections of the true stress-strain curves in the plastic region in Fig. 3.

Fig. 8, derived from Figs. 5, 6 and 7, and the observed spacings, indicates the relationship between the interlamellar spacing of the pearlite and all the properties measured. It will be noted that the plot of strength properties against the logarithm of the interlamellar spacing also yields a straight line. This requires that a certain fractional change in interlamellar spacing produce a certain absolute change in property; that is $dP = a \, dx/x$, where P is the property, x is the spacing, and a is a proportionality constant. This integrates to $P = 2.3 \, a \cdot \log x + C$.

The spacings indicated in Fig. 8 below 300 Ångström units were obtained by extrapolation and probably have no physical significance. However, the remarkable straightness of the property curves, when plotted against the reaction temperature, strongly tempts one to speculate on the possibility that the spacing is that indicated by the extrapolation, and that in the bainite structure what appear to be single phase particles are really aggregates. Going still further, one might assume them to be lamellar aggregates. It must be remembered that the spacing measurements, upon which

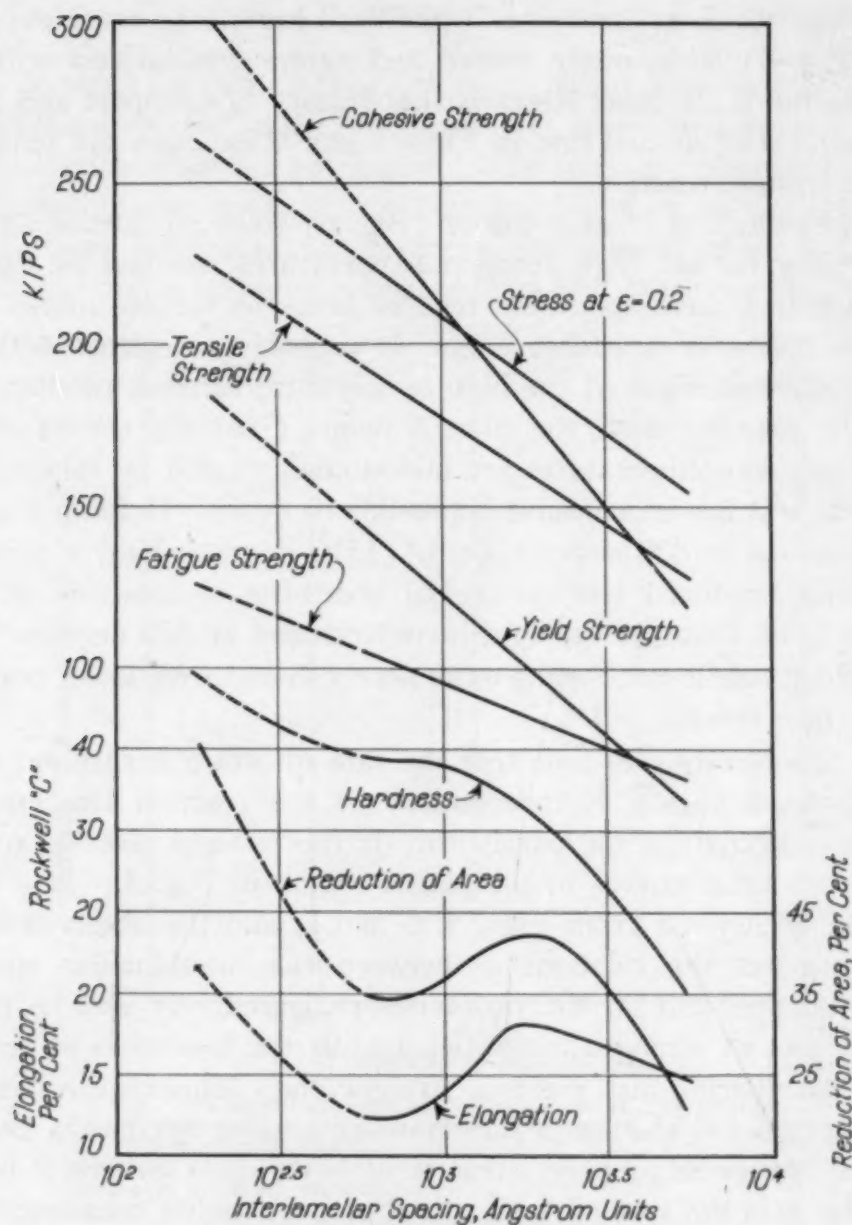


Fig. 8—Mechanical Properties Versus Interlamellar Spacing. Each property is plotted against the logarithm of the interlamellar spacing. Where the lines are dashed the spacing values were obtained by extrapolation.

the extrapolation is based, need to be verified. Perhaps all that can be said with certainty is that in an aggregate of ferrite and cementite, the strength of the aggregate increases as the degree of dispersion becomes greater, and that in the range of dispersions which can be studied there is a linear relationship between the logarithm of the interlamellar spacing and both the resistance to deformation and cohesive strength.

It is to be emphasized that the specimens are clearly entirely pearlitic only above about 600 degrees Cent., and even in this range there may be some spheroidization, as there certainly is above 675 degrees Cent. Between 600 degrees Cent. and 500 degrees Cent. the structure appears to be pearlitic, although it could not be entirely resolved. At 475 degrees Cent., bainite appears, mixed with pearlite. At 430 degrees Cent., no resolvable pearlite was found. At 370 degrees Cent., the mixed acicular structure illustrated in Fig. 1-H was obtained.

The size of the test specimens used (0.26-inch diameter) naturally raises a question concerning the effect of this rather large size on the structures obtained. To check this, disks 0.06 inch thick and about 0.375 inch in diameter were given the same treatment as the larger specimens, except for faster cooling to room temperature after the reaction. The structures obtained were, as nearly as could be estimated, identical with the structures in the larger specimens, except that no pearlite was found at 475 degrees Cent. and bainite could be identified for the first time at 496 degrees Cent. These specimens indicated that the least reaction time was at 550 degrees Cent., which corresponds to the observations of Bain and his associates. (Ref. Davenport and Bain, loc. cit.) In spite, then, of the regular way in which properties change with the reaction temperature, some recalescence may have occurred during the "isothermal" reaction of the larger specimens; the pearlite observed in the larger specimens may have formed at a higher temperature than the one reported. The reaction products for strictly isothermal reactions have not as yet been definitely identified. It is not yet certain that pearlite can form at temperatures below that of the fastest reaction (the so-called knee of the Bain S-curve) (Ref. Mehl, loc. cit.); or, again, it may be that pearlite is nucleated on passing through the knee on cooling to the reaction temperature.

The authors are extending this work to cover eutectoid alloy steels and propose to work with hypo- and hypereutectoid steels, both plain carbon and alloy. In the alloy steels, with their slower reactions, less trouble should be experienced due to recalescence and high temperature nucleation. Some experiments not ready for publication indicate that results similar to the above may be expected.

SUMMARY

1. The interlamellar spacing of the pearlite produced in the isothermal decomposition of austenite in plain carbon eutectoid steel

is related to the temperature of the reaction by either (1), the expression $\ln x = \frac{a}{T} + c$, or (2), $\ln x = aT + c$, where x is the interlamellar distance, T is the absolute temperature of the reaction, and a and c are constants. The accuracy of the measurements does not permit a choice between these two expressions.

2. The following properties are straight line functions of the reaction temperature, decreasing as the reaction temperature increases: (a) tensile strength, (b) yield strength (stress to produce a permanent extension of 0.2 per cent), (c) stress to produce any effective deformation in excess of 20 per cent, (d) fatigue strength, and (e) technical cohesive strength.

3. This relationship is valid to temperatures below that at which pearlite can be recognized and breaks down only when it is suspected that some martensite is produced by the heat treatment used.

4. Rockwell C hardness is not a linear function of the reaction temperature.

5. The fatigue strength is half the tensile strength only for the high temperature reaction product.

6. The rate of strain hardening in the plastic range beyond the maximum load in the tensile test is a constant, independent of the reaction temperature.

7. Elongation and reduction of area in the tensile test do not vary in a linear way with reaction temperature; they exhibit a minimum about 100 degrees Cent. below the knee of the Bain S-curve and a maximum at the knee. The reduction of area is least at the highest reaction temperature.

8. It is observed that down to a reaction temperature of 500 degrees Cent. the properties listed under item number 2 above are linear functions of the logarithm of the pearlite spacing, within the limits of accuracy of pearlite spacing measurements.

9. The linear relationship between mechanical properties and the logarithm of the pearlite spacing requires that a certain fractional change in spacing produce a certain absolute change in resistance to deformation and cohesive strength over a wide range of pearlite spacings; that is, $dP = a \, dx/x$.

10. It is suggested that in the low temperature reaction products the spacing is that obtained by extrapolation from the higher temperature range in which it can be measured directly.

11. For the steel used, the following relations hold for temperatures of reaction from about 350 to 700 degrees Cent. (660 to 1290 degrees Fahr.) if spheroidization is avoided.

Fatigue Strength (psi)	= 178,000 - 160T = 215,000 - 40,000 log X
Yield Strength (psi)	= 302,000 - 350T = 383,000 - 86,700 log X
Tensile Strength (psi)	= 316,000 - 270T = 377,000 - 66,600 log X
Cohesive Strength (psi)	= 474,000 - 510T = 587,000 - 124,800 log X

T is temperature of reaction in degrees Cent. X is interlamellar spacing in Ångström units ($\text{mm.} \times 10^{-7}$).

ACKNOWLEDGMENT

The authors wish to acknowledge the assistance and support of Dr. R. F. Mehl, director of the Metals Research Laboratory, and Messrs. W. A. Johnson and F. C. Hull, graduate students at the Carnegie Institute of Technology. They wish especially to acknowledge the assistance of Dr. Cyril Wells of the Metals Research Laboratory, on whose advice they relied greatly. The Page Steel and Wire Company furnished the coil of wire rod.

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DISCUSSION

Written Discussion: By A. V. deForest, professor, Massachusetts Institute of Technology, Cambridge, Mass.

This paper constitutes one of the very finest pieces of correlation between metallographic structure and mechanical properties which has yet appeared.

In all too many previous works the structure variables were so various and uncontrolled, the mechanical tests so crude, and limited, that a complete picture of the field was not apparent. In this case, both sides of the problem have been adequately presented and the result is extraordinarily consistent.

A statement is made which deserves the attention of all students of strength of materials, namely that the stress-strain curve, the fatigue strength, and the cohesive strength, adequately describe the mechanical behavior of metals. I believe this statement covers a very large portion, if not the whole, of the field of mechanical strength, but a large amount of corroborative work will have to be done to prove this proposition. The cohesive strength of metals is one of the fundamental properties which has not been apparent from commonly used methods of test, but the method described in this paper, if it is proved to be dependable, will allow such measurements to be made in future investigations.

The fatigue crack method of measuring cohesive strength should some day be checked against the triple-stress method of Haigh and if results are concordant, or a deficiency between the two methods can be successfully explained, we may look forward to important progress in the rational treatment of mechanical stress problems. It may be that using the fatigue crack method a

P
—
A

simple — value of cohesive strength can be depended on but if there exists an indeterminate stress distribution over the uncracked area of the test piece, the maximum value of cleavage strength rather than the average may become important.

There is no need to point out the difficulty of recording a figure covering fatigue strength. Rotating-beam figures on polished specimens certainly provide concordant results, which probably, in themselves, are rather to be termed fatigue strength under special circumstances than a fundamental figure applicable to certain metallographically well defined structures. I trust that Dr. Gensamer will elaborate further on his excellent work in this field.

Written Discussion: By R. H. Heyer, junior metallurgist, The American Rolling Mill Co., Middletown, Ohio.

The authors are to be complimented on this presentation of the mechanical properties of hot quenched eutectoidal steel. Their results on other carbon and alloy steels are awaited with interest.

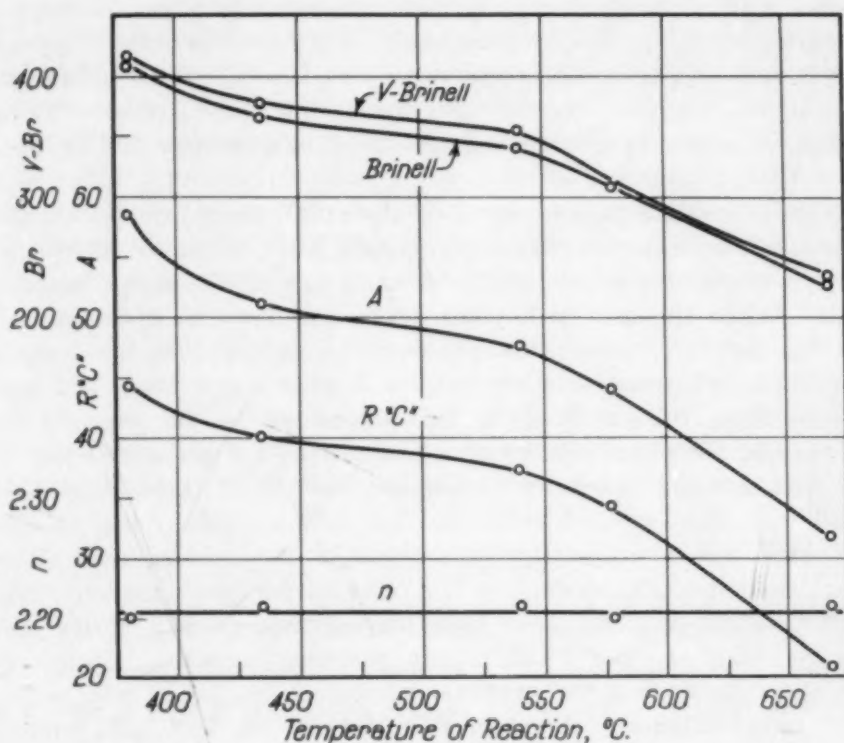
It is the writer's opinion that the value of the research would be greatly increased by the inclusion of properties of quenched and tempered samples. Comparison of the results in the hardness range over 35 Rockwell "C" would be of special significance.

The writer was impressed by the constancy of the strain hardening rate in the plastic range beyond the maximum load in the tensile test. The authors kindly furnished representative samples for the purpose of determining the Meyer hardness constants as a check on strain hardening rates. The results

of these and other hardness tests are summarized in the accompanying figure.

All hardness tests were made on polished transverse sections. A $\frac{1}{8}$ -inch steel ball was used in the Meyer hardness tests. Loads from 100 to 1200 pounds were applied in an Amsler testing machine.

The Meyer index "n," which is a measure of strain hardening capacity, is definitely constant throughout the range of reaction temperatures studied. The curve "A" represents the load in pounds required to make an impression whose diameter is $\frac{1}{10}$ the diameter of the ball. The Brinell curve was calcu-



lated from the Meyer hardness data for impression diameters equal to 0.35 times the ball diameter. The Vickers-Brinell data were obtained in a Vickers tester using the standard diamond pyramid indenter.

The Rockwell "C" curve is about 6 points higher than that given by the authors. Additional tests made on the curved cylindrical surfaces checked Fig. 6, therefore it seems probable that the authors' hardness tests were made in this manner.

Authors' Reply

The authors wish to thank Professor deForest and Mr. Heyer for their worthwhile discussions. In reply to Professor deForest, we should like to say that we are proceeding with tests intended to estimate whether or not our method for measuring cohesive strength compares with other methods. As to fatigue properties under other conditions than those prevailing in the ordinary fatigue test, it seems to the authors that the evidence is now quite good that the shear strain energy hypothesis predicts fatigue behavior under combined

stresses, but it is certainly true that much remains to be learned about the effect of the range of stress. It is our hope, however, that ultimately a simpler theory may be developed so that the fatigue stress as determined in the ordinary way will be adequate to describe the fatigue properties when taken in conjunction with cohesive strength measurements.

It is gratifying to learn that the Meyer method of studying the rate of work hardening confirms our conclusion concerning the constancy of the rate of work hardening in our samples. Mr. Heyer has done us a signal service in calling attention to the errors in hardness as originally reported and as appear in Fig. 6 of the preprint. We have re-determined the hardness on the same samples, checking Mr. Heyer exactly, and have corrected Fig. 6 in the paper as it will appear in the TRANSACTIONS. It may be remarked that our hardness figures are now in good agreement with those reported by Davenport and Bain in the paper referred to in our bibliography and in their frequently published S-curve.

The authors would like to say that since this paper was submitted, they have done further experimental work which leads them to believe that in some instances the whole specimen did *not* react at the temperature of the lead bath. When the lead bath temperature was between approximately 500 and 600 degrees Cent. part of the specimen reacted at a higher temperature. How much of the specimen reacted at this higher temperature and what the temperature of the reaction really is the authors are not yet prepared to state. They hope that the error introduced is not serious but are afraid that it may be. The fact that in this temperature range they check the hardness values of Davenport and Bain coupled with the fact that a smaller size of specimen gave identical tensile properties leads them to the opinion that all of the isothermal decomposition studies so far reported for plain carbon steels may be in error in this respect. They have learned that the size of the specimen is not as important as other details of the quenching technique. If future work confirms these suspicions, the graphs reported in this paper will need to be revised to indicate the actual temperature and not the lead bath temperature. This revision will not in any way modify the relationships developed between the properties and the interlamellar spacing of the pearlite, for those were measured directly. The only important modifications of the conclusions stated in the summary of the paper will be in the formulas for properties as a function of the reaction temperature in item 11 of the summary and a modification of item 10, which will probably have to be withdrawn, for it is based entirely on the continuity of the lines in Fig. 5. These lines will probably show a break when plotted against the true reaction temperature. The work now in progress on alloy steels, which are slower reacting, should be of interest in this connection.

HARDNESS CONVERSION RELATIONS FOR HARDENED STEELS

BY HOWARD SCOTT AND T. H. GRAY

Abstract

Previous work having justified the use of conversion relations between Rockwell and other penetration hardness tests, conversion relations were prepared from carefully verified data. All hardness values were compared directly with the diamond pyramid scale which offers the most satisfactory reference standard.

The conversion relation between Rockwell "C" and diamond pyramid hardness for steel is materially different from that for sintered carbides at comparable hardness so separate tables are given for each material. Both relations are accurately expressed by a simple empirical equation.

The physical property responsible for change in the conversion relation with test metal composition is shown to be the elastic modulus. By introduction of a term for elastic modulus the equation was so modified as to represent nonferrous metals as well as steel and carbides.

INTRODUCTION

BECAUSE of wide discrepancies between published hardness conversion charts and the need for reliable data, the writers were asked to prepare an accurate table of relations between the important hardness scales by the Inspection and Standards Departments of Westinghouse Electric and Manufacturing Company if such relations were found to be justified. This required experimental work, one phase of which has been completed and reported.¹

The experimental work reported was confined largely to a comparison of the Rockwell "C" (Rc) and diamond pyramid hardness (DPH) tests because of the major importance of the former and the fact that the latter test provides our best available reference standard. The diamond pyramid hardness tests were made on the

¹H. Scott and T. H. Gray, "Relation Between the Rockwell "C" and Diamond Pyramid Hardness Scales," TRANSACTIONS, American Society for Metals, Vol. 27, 1939, p. 363.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, Howard Scott is section engineer, and T. H. Gray, research department, Westinghouse Electric & Mfg. Co., E. Pittsburgh, Pa. Manuscript received March 24, 1939.

Vickers machine. Conversion relations between these and related tests were found to be fully justified when the testing conditions are adequately stated and the metal tested is hardened steel. There was no effect due to steel composition except in one particular case mentioned under testing precautions.

The major causes of discrepancies between various conversion charts may be attributed to one or more of the following factors:

- (1) Inaccuracy of test machine
- (2) Use of metals differing widely in elastic constants
- (3) Decarburized skin on test specimens
- (4) Brinell balls of different hardness

Avoiding these and other possible sources of error, the results of our basic relation of R_c to DPH has been combined with other research data and with carefully selected published observations to produce the desired conversion tables presented here.

FORM OF CONVERSION TABLE

The suggestion has been made by the Inspection Department that all hardness readings be expressed by a single scale value no matter what instrument was used for the particular test. This is certainly the only reasonable way in which to meet the confusing situation created by the great variety of hardness tests and scales in common use. That it can properly be done under moderate restrictions is established by the paper quoted. It remains then only to put verified data in the most convenient form for general use.

Conversion relations are most accurately and compactly given in the form of curves, but curves are not acceptable to the average user. Consequently the custom is to put them in the form of tables. This practice, however, is objectionable for our purpose because it does not permit use of integral values in more than one column. For this reason also accurate interpolations are difficult. This objection has been overcome by the device of making a scalar table, that is, entering integral values only at their exact value on a vertical reference scale. Thus exact conversion by interpolation is possible between any pair of hardness scales simply by placing a straight edge across the sheet.

Such a scalar table is given in Fig. 1. The lines under each number give the exact value of DPH on the equally spaced scale at both edges. The diamond pyramid scale is so used for a reference standard because it is least distorted of any hardness scale. It

DIAMOND PYRAMID HARDNESS (50 KG. LOAD) DPH	ROCKWELL HARDNESS					SCLEROSCOPE HARDNESS	MONOTRON LOAD VALUE 9 DIVISION PENETRATION MON.	BRINELL HARD- NESS HULTGREN 10 MM. BALL 3000 KG. LOAD		DIAMOND PYRAMID HARDNESS (50 KG. LOAD) DPH
	C SCALE - 150 KG. Rc	A SCALE - 60 KG. Ra	SUPERFICIAL							
			15-N SCALE 15-N	30-N SCALE 30-N	45-N SCALE 45-N					
								DIA. IN M.M. M.M.	HARDNESS NUMBER BHN	
1000	70.5					103	116		1000	
	70	86.5		86	77.5	102	114			
			94				112			
	69	86		85.5	77	100	110			
950			93.5	85	76.5	98	108		950	
	68	85.5		84.5	76		106			
				84	75	96	104			
900	67	85	93			94	102		900	
				83	74		98			
	66	84.5	92.5		73	92	96			
850		84		82	72	90	94		850	
	65		92				92			
	64	83.5		81	71	88	92			
800		83	91.5	80	70	86	90		800	
	63						88			
	62	82	91	79	69	84	86			
	61		90.5	78	68		84			
750		81	90	77	66	82	82		750	
	60			76	65		80	2.50	603	
	59	80	89	75	64	78	78	2.525	592	
700							76		700	
	58	79		74	62	76	74	2.55	576	
			88	73	61	74	72			
650	54	78		72	60	72	70	2.60	555	
			87	71	59		68	2.65	534	
	52	77		70	58	70	66	2.70	514	
600			86	69	56	68	64	2.75	495	
	50	76	85	68	55	66	62	2.80	477	
							60	2.85	460	
550	48	75		66	52	64	58	2.90	444	
	46	74	84	65	51	62	56			
			83	64	50		54	2.95	429	
500	44	73		62	48	60	54	3.00	415	
	42	72	82	62	46	58	52	3.05	401	
							50	3.10	388	
450	40	71	81	60	44	56	48	3.15	375	
	38	70	80	58	42	54	46	3.20	363	
	36	69	79	56	40	52	44	3.30	341	
400									400	
	34	68	78	54	38	50	42	3.40	321	
	32	67	77	52	36	48	40	3.50	302	
350									350	
	30	66	76	50	32	46	38	3.60	285	
		64	74	48	30	44	36	3.70	269	
300									300	
	25	62	72	46	28	42	34	3.80	255	
							32	3.90	241	
250									250	
	20	60	70	44	26	40	30	4.00	229	
				42	20	38	28	4.10	217	
200		58	68			36	26	4.20	207	
						34				
						32				

Fig. 1—Scalar Hardness Conversion Chart for Hardened Steels.

covers a very wide range of hardness, and the testing instrument used can be easily calibrated from standards of length and weight.

There still remains the need for data in tabular form from which exact relations between any two scales can be obtained, as for example to provide a diamond pyramid hardness scale for a particular hardness tester. To meet this need Table I was prepared from the same data as the scalar table. The source and reliability of these data are discussed in the next section.

SOURCE AND RELIABILITY OF DATA

The data for the most important relation, that of Rockwell "C" to DPH, was taken in the manner shown in the paper cited. A new Vickers hardness tester with an accurately calibrated load was used to obtain the diamond pyramid hardness. As an additional check on the Vickers machine used for the DPH tests a proof ring was made to determine the load actually applied to the work through the penetrator. This ring was carefully calibrated by dead-weight loading and gave the following results when applied to the Vickers machine:

Load in Kilograms		Per Cent Error
Nominal	Observed	
50	49.75	-0.5
30	30.15	+0.5
10	9.95	-0.5
5	4.95	-1.0

The standard of Rockwell hardness is, of course, the values on test blocks certified by the manufacturer of the machine. Our DPH observations on such blocks established the curves, Figs. 8 and 9, from which the tabulated values of Rc were taken.² This relation agrees very well with the recently issued Chart 38 of the Wilson Mechanical Instrument Company, maker of the Rockwell machine, except at the highest values. The difference is less than 1.0 per cent for values under 875 DPH. Our observations terminated at 350 DPH, but in view of the good agreement were extended to 200 DPH by means of Chart 38.

At high hardness the difference between the Wilson values and those obtained here is unduly large and not explained alone by the fact that Wilson DPH tests were made with a 10-kilogram load and

²These curves give slightly lower Rockwell values at a particular high diamond pyramid number than those given in our previous paper. This effect is probably due to better optical definition in the new Vickers machine used for the present work because no appreciable loading or measuring error could be detected in either instrument.

consequently are less accurate than those named here with a 50-kilogram load. Another cause for the discrepancy was suspected, namely, the use of sintered carbide test pieces to obtain the Wilson hardness relation at very high values. This suspicion was confirmed by tests made here on carbides and described in the following section. The hardness conversion relation for carbides is found to be materially different from that for steel so a separate conversion table is given. The main table consequently applies only to steels and does not extend into the hardness range of carbides.

The other Rockwell scales using a diamond cone penetrator, namely, "A" for standard Rockwell machine and 15-N, 30-N, and 45-N for the superficial machine,³ presented no need for laboratory work here. From the design of the machine these scales must be linear with reference to the "C" scale or nearly so. Their displacement relative to the "C" scale has been determined and is undoubtedly maintained with great care by the manufacturer of this machine, since it is essential to the integrity of his product. Their relation, nevertheless, has been checked and confirmed here over the range 825 to 1000 DPH. Accordingly, the inter-relations between the Rockwell scales given here were taken from Wilson Chart 38. Thus the relations of all Rockwell scales to diamond pyramid hardness are established through the "C" scale though confirmed directly over part of the hardness range.

Accurate Brinell testing can be done only on metals under about 600 BHN (680 DPH) even when a work hardened Hultgren test ball is used as in the results presented here. At higher hardness the ball deforms permanently and gives spurious readings. Observations made with standard Brinell balls may give lower readings and consequently require a different conversion relation. At the other extreme, with use of a diamond test ball, the Brinell values are practically identical with DPH readings. Since many hardness conversion charts have used Brinell hardness as a reference standard, it is easy to see how large discrepancies within the chart or between charts may originate alone from failure to specify the type of test ball used.

To specify more completely the test balls used here, they were tested for DPH on a Vickers machine using both a 50- and 10-kilogram load. Both on the spherical surface and on flat ground spots

³The essential difference between these machines is that the scale divisions of the standard Rockwell tester correspond to 0.0002 centimeter depth of impression and on the superficial to 0.0001 centimeter.

Table I
Hardness Conversion Table for Hardened Steels

Rockwell Hardness										
Diamond Pyramid Hard- ness (50 Kg.)	"C" Scale—150 Kg.	"A" Scale—60 Kg.	Superficial			Scleroscope Hardness	Monotron Load Scale Value (9 Divisions)	Brinell Hardness 10 mm. Hultgren Ball 3000 Kg.		Diamond Pyramid Hard- ness (50 Kg.)
DPH	C	A	15-N Scale	30-N Scale	45-N Scale	Scl.	Mon.	m.m.	BHN	DPH
1060	71.0	87.0	94.5	86.8	78.5	105	119	1060
1040	70.6	86.9	94.4	86.5	78.2	104	117	1040
1020	70.3	86.7	94.2	86.2	77.8	102.5	115	1020
1000	69.9	86.5	94.1	85.9	77.5	101	113	1000
980	69.5	86.3	93.9	85.5	77.0	100	110.5	980
960	69.0	86.0	93.7	85.1	76.5	98.5	108	960
940	68.5	85.7	93.5	84.7	76.0	97	106	940
920	68.0	85.4	93.3	84.3	75.5	96	104	920
900	67.4	85.2	93.1	83.9	75.0	94.5	102	900
880	66.8	84.9	92.9	83.4	74.0	93	100	880
860	66.4	84.6	92.7	82.9	73.5	92	98	860
840	65.6	84.3	92.4	82.3	72.8	90.5	96	840
820	64.8	83.9	92.1	81.7	72.0	89	94	820
800	64.1	83.5	91.8	81.0	71.2	87.5	92	800
780	63.3	83.0	91.4	80.3	70.2	86	89.5	780
760	62.5	82.5	91.1	79.6	69.2	85	87.5	760
740	61.6	82.0	90.7	78.9	68.2	83.5	85	740
720	60.6	81.5	90.3	78.1	67.2	82	83	720
700	59.6	81.0	89.9	77.2	66.0	80.5	81	2.485	612	700
680	58.6	80.5	89.5	76.3	65.0	79	79	2.500	603	680
660	57.6	80.0	89.1	75.4	64.0	77.5	77	2.520	592	660
640	56.6	79.5	88.7	74.5	62.8	76	75	2.545	580	640
620	55.6	78.9	88.2	73.5	61.5	74.5	73	2.570	567	620
600	54.5	78.3	87.7	72.6	60.2	73	71	2.600	553	600
580	53.4	77.7	87.2	71.6	59.0	71.5	68.5	2.635	538	580
560	52.3	77.1	86.7	70.5	57.8	70.0	66.5	2.675	522	560
540	51.1	76.4	86.1	69.4	56.5	68.5	64.5	2.720	506	540
520	49.8	75.7	85.5	68.3	55.0	67	62.5	2.770	490	520
500	48.5	75.0	84.8	67.1	53.2	65.5	60.5	2.820	473	500
480	47.1	74.3	84.1	65.8	51.5	64	58	2.875	456	480
460	45.6	73.5	83.4	64.4	49.8	62	56	2.930	439	460
440	44.0	72.7	82.6	63.0	48.0	60	54	2.985	422	440
420	42.3	71.8	81.7	61.5	46.0	58	52	3.045	404	420
400	40.5	70.9	80.8	59.9	43.8	56	50	3.115	386	400
380	38.5	69.9	79.8	58.2	41.5	54	48	3.195	362	380
360	36.5	68.8	78.7	56.3	39.2	52	46	3.285	344	360
340	34.3	67.7	77.5	54.3	36.5	50	43.5	3.375	326	340
320	32.0	66.5	76.2	52.2	33.8	47.5	41	3.465	308	320
300	29.5	65.2	74.8	50.0	30.8	45	38.5	3.560	294	300
280	26.9	63.8	73.3	47.7	27.8	42.5	36	3.660	275	280
260	24.0	62.3	71.7	45.2	24.5	40	33.5	3.780	256	260
240	21.0	60.7	69.9	42.1	21.0	37.5	31	3.920	237	240
220	35	28	4.080	218	220
200	32	25	4.270	199	200

the hardness was 970 DPH. Values on standard Brinell balls were 920 to 940 DPH.

The BHN-DPH relation used here is again that of Wilson Chart 38. It has been confirmed by tests made here and plotted as points in Fig. 2, the solid curve of which represents the Wilson

data. Work done here at an earlier date is also confirmatory of this.

Scleroscope hardness is referable only to the standards of the maker of this instrument. Accordingly, DPH tests were made here on his test blocks and the results plotted in Fig. 3. This curve supplied the values given in the scalar table and in Table I. The rela-

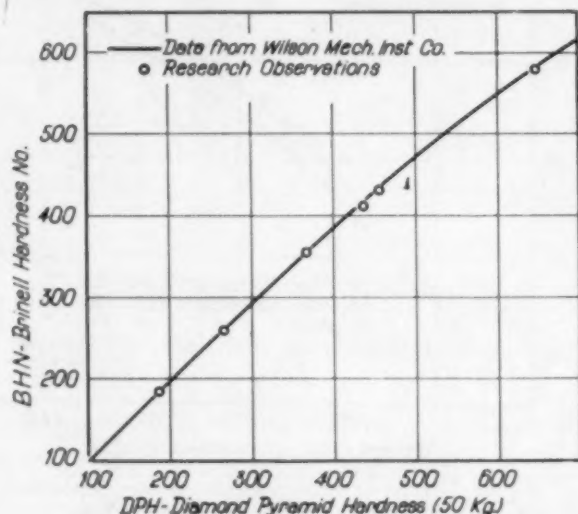


Fig. 2—Diamond Pyramid Hardness-Brinell Hardness Relations.

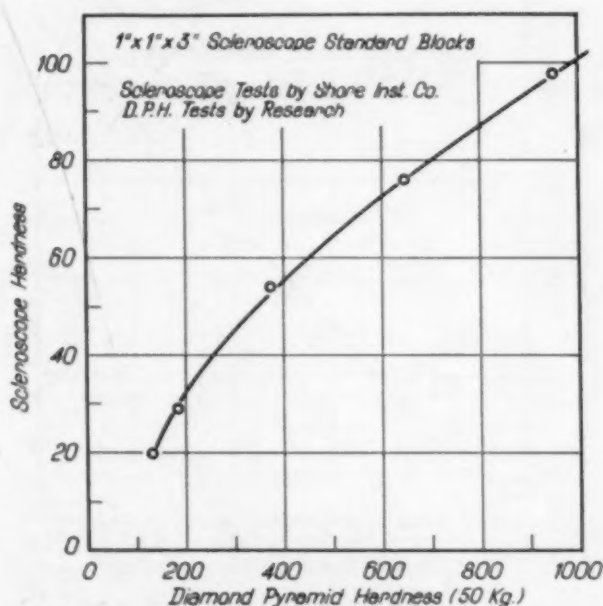


Fig. 3—Diamond Pyramid-Scleroscope Hardness Relations.

tion is applicable only to fairly large test pieces. Smaller blocks than the ones used which were 1-inch square give a lower scleroscope value at a particular DPH as shown in Fig. 4. It is probable that the elastic modulus of the test metal affects the relation also.

The hardness tests presented so far are those most commonly used. Another test used to some extent is the Monotron. This test uses a diamond ball penetrator having a diameter of 0.75 mm. It

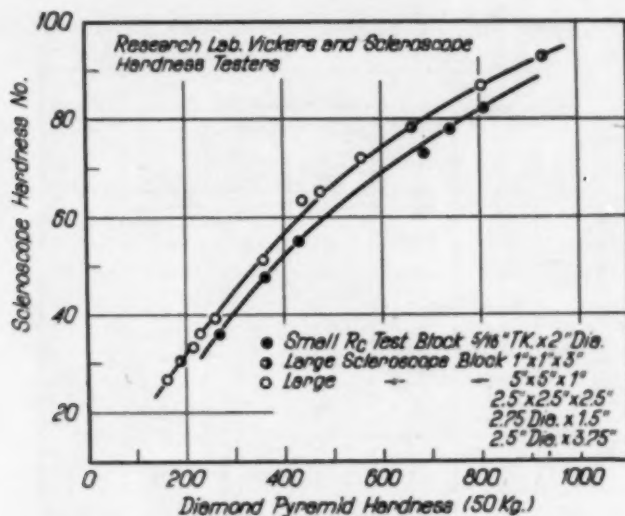


Fig. 4—Diamond Pyramid-Scleroscope Hardness Relation for Large and Small Size Test Blocks.

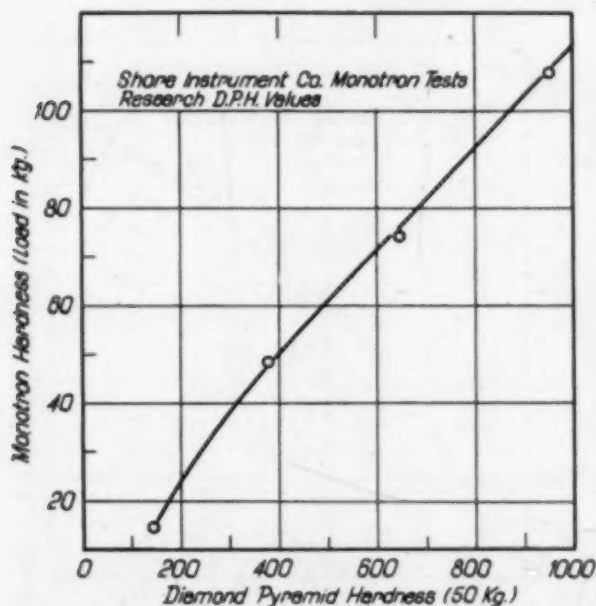


Fig. 5—Monotron-Diamond Pyramid Hardness Relations.

is pressed into the work to a constant depth indicated by a scale reading of 9.0 divisions which corresponds to an impression area of 0.1070 square millimeter, and the load is then read on another scale. Load readings made by the Shore Instrument Company on our sclero-

Table II
Monotron-Diamond Pyramid Hardness Relations

DPH (50 Kg.)	Monotron Load Value, P.	Calculated DPH	
		P	P-6
		A	A
1000	113.0	1056	1000
900	102.0	953	897
800	91.5	855	799
700	81.0	757	701
600	70.5	654	603
500	60.0	561	504
400	50	467	406
300	38.5	360	304
200	25	233	178

DPH and P values from Fig. 5. $A = 0.1070 \text{ mm.}^2$ for scale reading of 9 divisions which corresponds to 0.045 mm. depth.

scope test blocks are plotted against DPH readings made here in Fig. 5. The Monotron column in Table I and Fig. 1 is taken from this curve.

Since the Monotron test impression is always of the same depth and presumably of the same area, the load reading divided by the theoretical impression area should give a hardness number of the same dimensions and magnitude as the DPH. Actually the DPH value so calculated is high, Table II. Subtracting an empirical factor 6 from the load permits, however, an accurate calculation of DPH from the load reading over the range 300 to 1000 DPH as shown in the same table.

CONVERSION RELATIONS FOR CARBIDES

Sintered carbide tools differ markedly from hardened steels, both in elastic modulus and hardness. Their elastic modulus is

Table III
Hardness Tests on Firthite Sintered Carbides

No.	Diamond Pyramid Hardness		Wilson Chart		Rockwell Tests by Wilson Mech. Instr. Company				
	Research Lab.		No. 38	"C"	"D"	"A"	15-N	30-N	45-N
	50 Kg.	10 Kg.	10 Kg.	150 Kg.	100 Kg.	60 Kg.	15 Kg.	30 Kg.	45 Kg.
10	936	959	854	66.0	75.2	84.7	92.8	82.7	73.6
9	1037	1078	1571	69.2	78.2	86.8	94.1	86.2	77.6
8	1071	1132	1602	69.8	77.9	86.2	93.3	85.0	76.8
1	1249	1303	1641	73.0	81.1	88.6	95.0	88.1	80.8
6	1442	1491	1664	76.2	83.6	90.2	96.3	89.8	83.8
3	1505	1516	1695	76.6	83.8	90.3	96.1	90.0	84.5
5	1527	1570	1018	77.1	84.1	90.7	96.3	90.6	84.9
7	1540	1580	1058	77.4	84.0	90.5	96.2	90.0	83.9
4	1553	1599	1323	77.8	84.4	91.1	96.6	91.0	85.6
2	1691	1760	1865	80.0	86.1	92.2	96.9	91.8	87.5
11	1748	1771	1904	80.5	86.1	92.2	96.9	92.0	87.6

over 2.5 times that of steel, and their hardness range from 900 to 1700 DPH, the maximum for steel being barely over 1000. For this reason it was suspected that a hardness conversion relation between Rockwell and diamond pyramid hardness for one type metal would be materially different from that for the other.

To test this possibility, samples of sintered carbides were obtained from the Firth-Sterling Steel Company. These pieces had

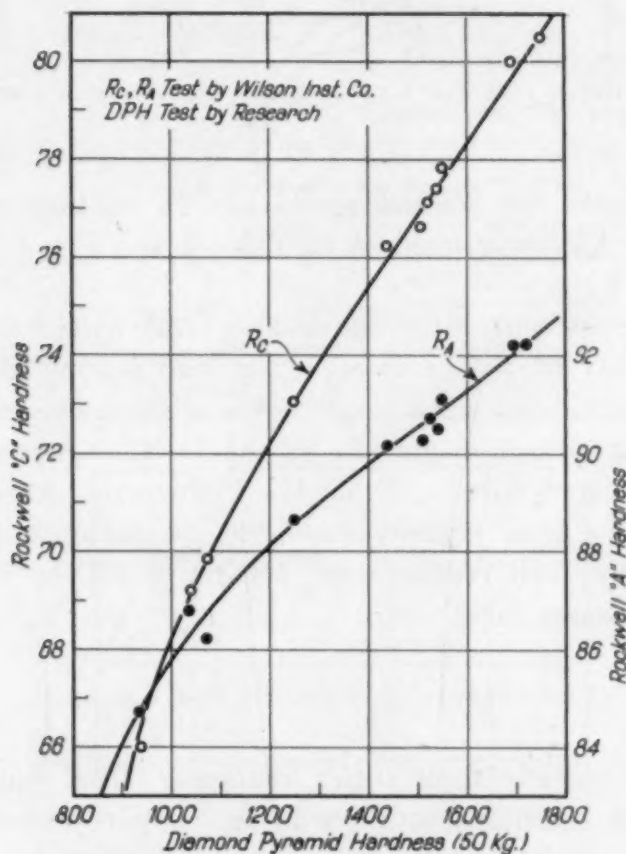


Fig. 6—Hardness Relations DPH vs. R_c , R_A for Sintered Carbides.

already been tested by the Wilson Mechanical Instrument Company for Rockwell "C", "D", and "A" scale hardness and for superficial hardness with 15-, 30-, and 45-kilogram loads. Their observations, together with our DPH tests, are given in Table III.

Values of R_c and R_A , the Rockwell tests most used on this material, are plotted against DPH in Fig. 6. The "A" scale is more commonly used because the lighter load reduces danger of chipping the diamond. From this curve were taken the values for our hardness conversion table for carbides, Table IV.

Table IV
Hardness Conversion Table for Sintered Carbides

Diamond Pyramid Hardness 50 Kg. Load	Rockwell A-Scale 60 Kg. Load Brale Penetrator	Rockwell C-Scale 150 Kg. Load Brale Penetrator
1750	92.4	80.5
1700	92.0	79.8
1650	91.7	79.2
1600	91.3	78.4
1550	90.9	77.7
1500	90.5	77.0
1450	90.1	76.2
1400	89.7	75.4
1350	89.3	74.6
1300	88.9	73.8
1250	88.5	73.0
1200	88.1	72.2
1150	87.6	71.3
1100	87.0	70.4
1050	86.4	69.4
1000	85.7	68.2
950	85.0	66.6
900	84.0	64.6
850	82.8	...

Going back now to the comparison of the conversion relations for steel and for sintered carbides, it may be seen from Fig. 7 that the Rc value at a particular DPH value in the region of overlap is about 2 points higher for steel than for carbides. The Wilson

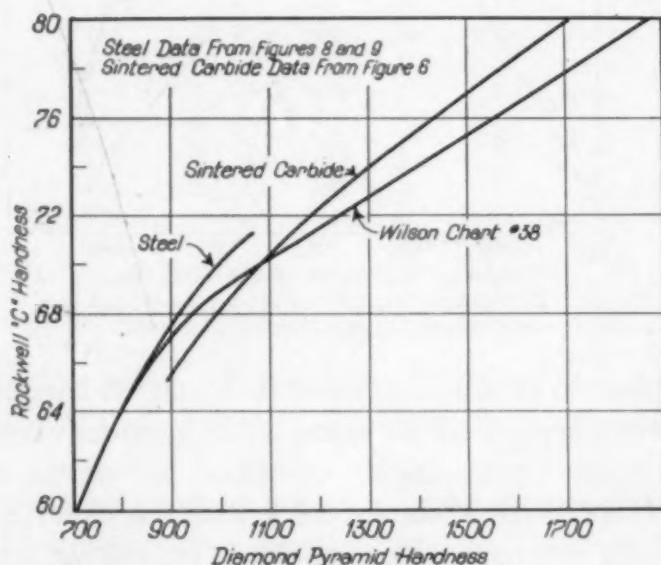


Fig. 7—Comparison of Diamond Pyramid-Rockwell "C" Relations at High Hardnesses.

Chart, however, does not distinguish between these materials and gives a curve which, with increasing hardness, gradually moves away from our curve for steel, cutting across our curve for carbides and falling much below the latter in the high hardness range.

Lack of agreement at the low end of the range is, of course, due to taking an average of the values on both materials. At the high end of the range it is due in part at least to use of a 10-kilogram load by the Wilson Company. The impression then is very small and cannot be accurately measured. Our observations of DPH on carbides with a 10-kilogram load also are high relative to those with a 50-kilogram load, Table III.

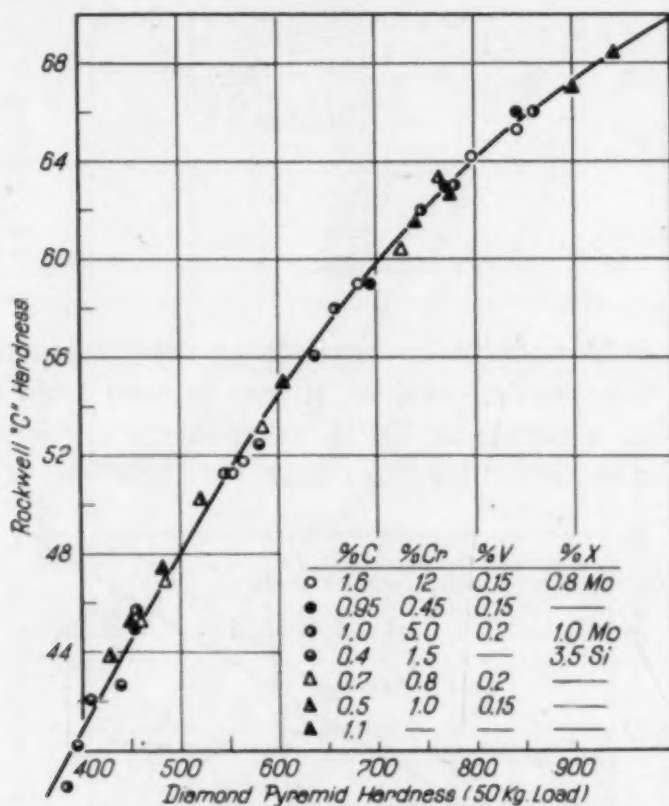


Fig. 8—Relation Between Rockwell "C" and Diamond Pyramid Hardness for Chromium Steels.

An explanation of the difference in Rockwell hardness between steel and sintered carbide of the same DPH is evident from a consideration of elastic "spring-back" described as elastic recovery in our previous paper. It is shown there that the metal at the center of the Rockwell impression springs back on release of the major load by an appreciable amount, 0.0005 to 0.0008 inch for steel with 150-kilogram load. The degree of spring-back will, of course, depend on the elastic modulus of the metal tested being greater for steel ($E = 30 \times 10^6$ pounds per square inch) than for sintered carbides ($E = 79 \times 10^6$ pounds per square inch). If, then, we can measure hardness as resistance to penetration by some means

uninfluenced by the elastic modulus, the Rockwell hardness number at a particular value by this test will be higher for steel than for sintered carbides. Because it uses a dimension in the plane of the test metal surface as a measure of hardness, the DPH test is probably affected little if any by the elastic modulus of the test metal. There should then be a difference in Rockwell hardness between steel and

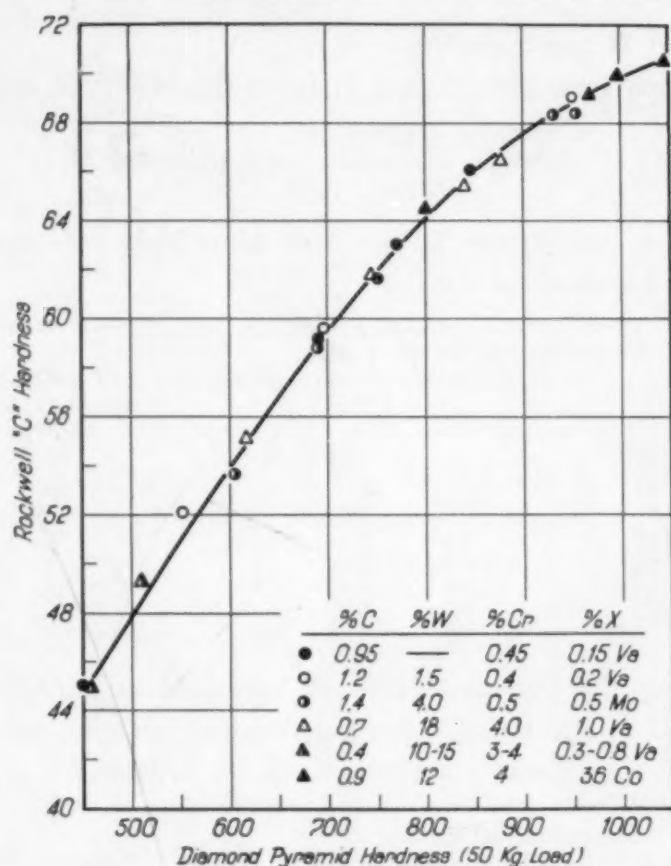


Fig. 9—Relation Between Rockwell "C" and Diamond Pyramid Hardness for Tungsten Steel.

tungsten carbide pieces of the same DPH as was observed, and the direction of the observed difference is that required by this analysis.

The anticipation of an effect due to elastic modulus and its experimental demonstration suggests that an equation might be developed which would contain a term for elastic modulus. Such an equation would provide an accurate means for computing DPH from Rockwell tests on any metal of known elastic modulus. A complicated equation has been derived by Petrenko⁴ on strictly theoretical grounds. It agrees with experiment fairly well as shown in

⁴Bureau of Standards Journal of Research 5:19 (1930).

our previous paper only when corrected for elastic recovery and then not close enough for practical use. The experimental relation between Rc and DPH, however, can be accurately represented by a simple empirical equation, namely:

$$DPH = \frac{A}{(B - Rc)^2} \quad (1)$$

where A and B are constants.

Evaluating constants A and B from the observed data on steel:

$$DPH = \frac{2.430 \times 10^6}{(119.0 - Rc)^2} \text{ for steel only}$$

The following comparison shows how accurately this equation represents the experimental data:

Experimental Values on Steel for :		Calculated Value of DPH
Rc	DPH	
71	1063	1056
68	922	934
65	825	833
60	708	699
55	609	594
50	523	512
40	393	390
30	302	307

Revaluing the constant B to fit the data on sintered carbides gave $B = 117.5$, and equally good agreement between calculated and experimental observations was obtained as follows:

Experimental Values on Sintered Carbides for :		Calculated Values of DPH
Rc	DPH	
80	1715	1744
78	1560	1570
76	1435	1420
74	1310	1293
72	1195	1182
70	1093	1084
68	995	999
66	925	923

The foregoing observations suggest that equation (1) can be modified to express the effect of elastic modulus simply by substituting $C + D/E$ for B where C and D are constants and E the elastic modulus of the metal tested. Evaluating C and D from the values of B for steel of $E = 30 \times 10^6$ pounds per square inch and for sintered carbide of $E = 79 \times 10^6$ pounds per square inch:

$$\text{DPH} = \frac{2.430 \times 10^6}{(116.33 + \frac{80.1 \times 10^6}{E} - \text{Rc})^2}$$

assuming the effect of elastic modulus to be linear.

Using the preceding equation, the DPH of copper and aluminum was calculated from Rc test values and compares with the observed values as follows:

Material Tested	Elastic Modulus	Observed Values of Rc	Observed Values of DPH	Calculated Value of DPH
2¼% Be-Copper	17.5×10^6	40	368	372
Al (14ST)	10.3×10^6	— 1	153	155
Al (17ST)	10.3×10^6	—16	125	124

Table V

Depth of DPH, BHN and Rockwell Hardness Impressions in Thousandths of an Inch

DPH	Diamond Pyr. Hard.		Brinell Hard. BHN	Depth	Rockwell Hardness			
	50 Kg.	10 Kg.			"C"	"C"	15-N	45-N
1000	1.71	0.76			70.0	2.56	0.29	1.07
950	1.76	0.79			69.0	2.65	0.31	1.12
900	1.81	0.81			67.5	2.77	0.34	1.20
850	1.86	0.84			66.0	2.90	0.37	1.29
800	1.92	0.86			64.5	3.03	0.39	1.36
750	1.98	0.89			62.0	3.24	0.44	1.51
700	2.05	0.92			59.5	3.45	0.48	1.66
650	2.12	0.95	586	6.40	57.0	3.66	0.52	1.76
600	2.21	0.99	553	6.81	54.5	3.88	0.57	1.89
550	2.31	1.03	514	7.37	52.0	4.09	0.61	1.99
500	2.42	1.08	473	8.00	48.5	4.39	0.69	2.19
450	2.56	1.14	430	8.77	45.0	4.69	0.78	2.36
400	2.71	1.21	386	9.74	40.5	5.07	0.90	2.61
350	2.87	1.30	335	11.48	35.5	5.50	1.02	2.90
300	3.09	1.40	294	13.40	29.5	6.01	1.17	3.23
250	3.35	1.52	247	15.39	23.0	6.57	1.31	3.57
200	3.82	1.71	199	18.61	15.5	7.21	1.50	3.98

DPH and BHN depths are calculated under the assumption that the impression contour is identical with that of the penetrator. Rockwell depth values are, of course, observed directly being proportional to the hardness number. Monotron test depth is constant at 0.00177 inch for 9 division scale reading.

The agreement in these individual cases is sufficiently good to suggest that this form of equation may be used in the low hardness range where it can be applied to much greater advantage than here. Work in this field is not sufficiently advanced for reporting at this time.

TESTING PRECAUTIONS

Hardness testing of thin metals introduces the requirement that the depth of impression be small relative to the metal thickness.

Inaccuracy occurs if the impression depth exceeds 7-10 per cent of the metal thickness. Also, if the metal tested is a composite one containing two layers of different hardness, the metal thickness to be considered is that of the layer under test. Thickness often controls the choice of hardness test itself.

To provide information on depth of test impressions so that errors of this kind may be avoided, Table V was prepared. Multiplying any depth value by 10 gives a conservative minimum thickness of metal that can be accurately tested under the conditions represented by the value chosen. Of course, when there is no limitation on depth of impression, the maximum load recommended for the tester can be used with advantage in accuracy of the test.

Error is introduced also by making tests too close to the edge or to other test impressions. According to O'Neill⁵ the impression should be at least two diameters away from the edge and five diameters distant from any other impression.

With the Rockwell test a common source of error on thin metal is a depression in the anvil immediately under the penetrator. This condition causes bending or deformation of the strip and consequently a spurious reading. For the same reason thin wall tubing must be tested as a segment cut from the tube and with the convex side next to the anvil. Burrs and dirt on the bottom face of the test piece also introduce error in the Rockwell test by permitting unrecovered motion of the work. This source of error is even more important when using the Rockwell superficial tester than when using the standard machine because the sensitivity of the former is twice that of the latter. The machine should be checked daily with reference blocks.

Each hardness testing instrument has its own particular sources of error. The Brinell test possibly gives least trouble, but still the ball must be selected to a hardness standard, and distortion of the ball by use on hard metals must be avoided. With the Vickers test commonly used to evaluate DPH, the surface tested must have a good polish and be so flat that the impression is rectangular or very nearly so. A small level can be conveniently used with the instrument to assure that the face of the test piece is in the plane of the anvil.

Load variation, where convenient as in the Vickers test, can be used to detect thin layers of carburized or decarburized surface

⁵"Hardness of Metals and its Measurement."

metal by tests on the surface alone. If the DPH value increases with load, the surface is softer than the core metal which indicates decarburization, and if it decreases, carburization is indicated.

Superficial hardening as in shallow hardening steels of large size, say $\frac{3}{4}$ inch thick or more, introduces a discrepancy in the relation between Rockwell "C" and diamond pyramid hardness. The Rc value may be 1 to 2 points higher on such steels than that obtained by conversion from a DPH reading. This effect appears to be associated in some way with the fact that the hardened surface layer is under extremely high compressional stress which disappears as the hardness is reduced by tempering.⁶

The foregoing case is the only one in which composition of steel enters as a factor affecting the conversion relations given here. Even in this case the effect of composition is secondary in the sense that it appears from the absence of alloy content rather than from its presence. It affects only the relation between a test measuring depth of impression and one measuring diameter. The relations between tests of each type is unchanged, at least the change is relatively small on going from a deep to a shallow hardened steel of high hardness.

SUMMARY AND CONCLUSIONS

Finding that reliable conversion relations between Rockwell and diamond pyramid hardness test values can be obtained on heat treated steels independent of composition, hardness conversion tables for the major hardness tests were prepared from the best available data and some new data presented here.

The diamond pyramid hardness test, having the best characteristics of accepted tests for a reference standard, was used for the linear reference scale of a scalar conversion table. This table permits conversion from a test value by any one of the following hardness tests to any other:

- | | |
|------------------------------|-------------------------|
| (1) Diamond pyramid hardness | (6) Rockwell 45-N scale |
| (2) Rockwell "C" scale | (7) Scleroscope |
| (3) Rockwell "A" scale | (8) Monotron |
| (4) Rockwell 15-N scale | (9) Brinell |
| (5) Rockwell 30-N scale | |

⁶O. V. Greene—"Estimation of Internal Stress in Quenched Hollow Cylinders of Carbon Tool Steel," *TRANSACTIONS, American Society for Steel Treating*, Vol. 18, 1930, p. 369.

Tests on sintered carbides reported here show that the relation between Rockwell and diamond pyramid hardness is dependent on the elastic modulus of the metal tested so that a separate conversion table is required for metals differing materially in elastic modulus from that of steel. A case in point is the sintered carbides which have a very high elastic modulus and for which an independent conversion table is offered. The origin of this effect is shown to be in the elastic spring-back with load release under the Rockwell test.

A simple empirical equation expresses accurately the relation between Rockwell "C" and diamond pyramid test values on steels. With change of a single constant the same equation gives equally well the same relation for sintered carbides.

Adding a term to this equation for elastic modulus, the new constants were evaluated from the observations on steel and sintered carbides. Values then calculated for metals of low elastic modulus, copper and aluminum alloys, agreed very well with a few experimental observations.

The information presented here fully justifies the use of a single reference scale of hardness for reporting all penetration hardness test values no matter which testing instrument of the kinds considered is used.

DISCUSSION

Written Discussion: By Arthur R. Kommel, metallurgist, United Engineering and Foundry Co., Vandergrift, Pa.

A hardness conversion table for hardened steels which will actually permit the results of any one of the common hardness tests to be translated into the terms of any other common test is certainly useful information.

However, it is perhaps well to point out that certain limitations and peculiarities of one of the tests used, namely the Scleroscope test, are such as to seriously affect the validity of any conversion.

With the "C" type or direct-reading Scleroscope as is most generally used, the personal equation enters into the results to a remarkable degree.

Table I shows the results of Scleroscope hardness tests by seven experienced operators on three 3 x 1 x 1-inch blocks, using two instruments. Each value reported is the average value of ten readings on each block for each operator. For purposes of comparison the Rockwell "C" hardness of each block is given together with the converted Scleroscope hardness as taken from the authors' Table A.

The deviations from the average readings as given in the second part of Table A show that certain operators are consistently high and others consistently low in their readings. This trend has been confirmed by repeated tests on

some of the same operators. It will be noted that the range of deviation is such, with the two hardest blocks, that conversion of readings to DPH from the authors' table results in a discrepancy of 70 to 100 points DPH, certainly too large if a close hardness specification must be met.

Table A
Scleroscope Hardness Readings

Rockwell "C" Converted Shore (Authors' Table I) Operator	Block 1		Block 2		Block 3	
	Sclero. A	Sclero. B	Sclero. A	Sclero. B	Sclero. A	Sclero. B
1	94.0	97.5	65.8	64.8	47.7	46.6
2	95.9	97.1	65.3	66.0	46.9	46.4
3	91.9	94.4	64.1	63.8	46.4	46.0
4	95.1	96.5	67.6	64.7	48.0	48.0
5	94.0	97.1	66.3	64.1	47.6	46.3
6	90.6	94.6	66.8	63.9	47.4	46.8
7	89.5	94.0	61.5	63.4	45.4	45.7
Average:	93.0	95.9	65.9	64.4	47.1	46.5
Deviation from Average						
1	+1.0	+1.6	+0.6	+0.4	+0.6	+0.1
2	+2.9	+1.2	+0.1	+1.6	-0.2	-0.1
3	-1.1	-1.5	-1.1	-0.6	-0.7	-0.5
4	+2.1	+0.4	+2.4	+0.3	+0.9	+1.5
5	+1.0	+1.2	+1.1	-0.3	+0.5	-0.2
6	-2.4	-1.3	+1.6	-0.5	+0.3	+0.3
7	-3.5	-1.9	-3.7	-1.0	-1.7	-0.8
Range (Shore)	6.4	3.5	6.1	2.6	2.6	2.3
Range (Converted DPH) @ 100			@ 70		@ 20	

Furthermore, the differences in calibration between certain instruments of the same make are such as to further complicate matters.

Two instruments, both of which had been standardized by the makers, were tested by the same operator on the same set of manufacturer's test blocks with the following results:

Block Marking	Shore Hardness		Points Difference
	Scleroscope 1	Scleroscope 2	
17/19	20	18	2
48/50	50	45	5
94/96	94	91	3

This shows that even after standardization by the makers, discrepancies between instruments may occur.

In another instance, two instruments were compared by the same operator on a series of test blocks, using some of the blocks which had been furnished with each instrument, with the following results:

Block	Shore Hardness	
	Scleroscope 1	Scleroscope 2
Scleroscope 2	48/50	51
Scleroscope 1	61/63	67
Scleroscope 1	80/82	86
Scleroscope 2	94/96	96

In this case, each instrument checked within reasonable limits the blocks supplied with that instrument, but except on the softest block showed great discrepancies when tested on the blocks supplied with the other instrument. This indicates that variation in the calibration of the manufacturer's own test blocks is such that an irregular instrument may not even be suspected, and

such variation certainly detracts from the validity of the authors' conversion for Scleroscope hardness as they apparently used the values supplied by the manufacturer, without checking.

The data presented here show that enough difficulty is present in comparing Scleroscope hardnesses as determined by various operators or with various instruments, let alone converting these values to some other scale and further complicating matters.

Written Discussion: By E. K. Spring, metallurgist, Henry Disston & Sons, Inc., Philadelphia.

The preparation of comparisons between the various methods of determining hardness has probably produced more varied relationships than any other attempt at standardization. Individual plants have had their own charts and cases have come to light where hardness specifications in vogue in one shop could not be applied elsewhere without suitable interpretation. Such conditions are rapidly vanishing due largely to the more intensive efforts at standardization of the past few years.

The form of conversion chart employed is probably largely controlled by the individual's personal inclinations.

The scalar chart does permit of more accurate comparison of values between those determined with different instruments. For quick, rough checks the tabular form is perhaps quicker and more suitable for general shop work. For laboratory work where accuracy of comparison is of more moment than speed, the scalar table has a definite place.

Since the work under test generally fixes the type of test which may be applied, these conversion figures are desirable. However, the same reason lessens the need for conversion, it being, in the writer's mind, preferable that hardness specifications be written for and obtained on that type of machine most suitable for the work in hand. This assumes that all shops undertaking general testing be equipped with a full complement of hardness testing machines, which condition is known to be untrue.

A point too often slighted in hardness determinations is preparation of the specimen according to either standard or proper methods. For example, the amount of metal removed in preparing for the test should be in accordance with some accepted figure. Surface finish can account for erratic results. Thickness of pieces tested should be set to eliminate anvil effect.

With standardization of these points much can be done to bring this type of testing on a plane throughout the industry.

The writers of the paper are to be complimented on an excellent piece of work tending to bring this matter of testing, for hardness, to a point where we all know what the other fellow is talking about.

Written Discussion: By Hugh O'Neill, chief metallurgist, London Midland and Scottish Railway Co., England.

The authors' experimental investigation of the hardness conversion relations is a useful piece of work, and the scalar chart given in Fig. 1 is a very acceptable way of presenting the results obtained. Following upon a demand from industry, the British Standards Institution, in May 1937, commenced the preparation of hardness scale comparisons and issued its Table in August 1939

as B.S.S. 860-1939. Published results, and the charts prepared by many independent firms, were used in the compilation of this Table, and the figures were submitted to a large number of engineering concerns for comment and comparison before the final issue. It may be said that the values in B.S.S. 860-1939 represent agreed practical results for a very large range of users in Britain, and it is notable that the authors' Rockwell "C" and "B" conversions agree well with the B.S.I. comparisons up to DPH 640 and 540 respectively, beyond which they become higher than the B.S.I. values, the difference at DPH 1000 being about 3 per cent. The reason for this is not clear unless there has been some change in constructional details of the Rockwell machines over a considerable period of time.

The principles upon which the Rockwell, Vickers and Brinell testing systems depend are outlined in Table B, and the possibilities of a theoretical conversion of the three scales may be considered.

Table B
Principles of Tests

Type of Test	Indenting Tool	Angle of Indentation	Reading Observed	Hardness Value*
(1) "Direct Reading" Scales C, D & A.	120° diamond cone.	Constant.	Depth (approx.) of "recovered" indentation.	$H = k - h'$
Scales B, E, F, G & S.	Hard steel ball.	Varies with load.	Depth (approx.) of "recovered" indentation.	
(2) Diamond pyramid.	136° diamond pyramid.	Constant.	True diagonal of "recovered" indentation.	$H = \frac{k}{h^2}$
(3) Brinell ball.	Hard steel ball.	Varies with load.	True diameter of "recovered" indentation.	$H = \frac{k}{h}$

*h = depth of indentation including ridge effect.
h' = depth of indentation from original datum.

The Rockwell and pyramid type tests would approximately give geometrically similar indentations if the Rockwell cone had an included angle of 142 degrees instead of 120 degrees.⁷ As it is, the Rockwell indenter would yield mean pressure hardness values about 4 per cent greater than those of the standard diamond pyramid due to this cause. Such a difference is unimportant compared with the large discrepancies from the fundamental point of view which arise by the adoption in the Rockwell test of a particular depth measuring system. Thus for both these indenters the pressure hardness values should theoretically be independent of the testing load, provided that true measurements of the size of the indentation are actually registered by the measuring machine. It is well known that this constancy is generally obtained in the pyramid test, but it is found that variations occur in the pressure hardness results calculated from a Rockwell type depth reading machine (see steel A, Table C). This variation-with-load effect in the Rockwell is to be expected

⁷G. A. Hankins, *P. I. Mech. Eng.*, Vol. 2, 1926, p. 823.

under the circumstances, and has been demonstrated for instance by Batson.⁸ Furthermore, Table C indicates the discrepancies between the corresponding cone and pyramid pressure hardness values for a selection of specimens, the former having been calculated from the Rockwell machine reading on the assumption that it indicates the real depth of the impression.

Table C
Experimental Results

Specimen	Brinell No. H _B (steel ball)	Diamond Pyramid Hardness No. Testing load, kg.	H _D kg/sq. mm.	Diamond Cone Hardness No. from results of Rockwell Test	
				Nett Testing load, kg.	H _C = $\frac{L^*}{10.88h^2}$ kg/sq. mm.
Aluminum	22	10	20	50	34
Steel A (0.25 C)	149	60	145	50	483
		100	140	90	350
		120	143	140	281
Steel VRA (0.4 C)	185	50	185	140	435
Steel VRA (0.4 C cold-rolled)	216	50	226	140	565
Steel 100 S (0.9 C water quenched)	670	50	965	140	2960

*Load/conical area. 1 Rockwell unit = 0.002 mm. depth.

It is evident that whilst both methods involve resistance to indentation, no exact theoretical correlation can be said to exist between these two testing systems, and the only possibility is an empirical comparison scale.

Considering next the ball test, there is no theoretical reason why the Brinell Number of a metal should necessarily equal its pyramid number even in principle. The respective indentations made by the two tests need not always possess geometrical similarity, and the different work hardening capacities of different specimens are therefore involved. The same applies to the Brinell ball in relation to cone indenter tests, and whilst Krupkowski⁹ has shown that the ratio H_C/H_B is often about 1.2, it ranges between 1.0 and 1.34 and is higher the greater the work hardening capacity of the specimen.

Regarding the Monotron test, the authors state that the load reading divided by the constant theoretical impression area should give a hardness number of the same dimensions and magnitude as the DPH. It must be noted that the theoretical Monotron impression area does not take account of the indentation ridge effect, and that unrecovered indentations are measured, whilst the standard impression area of 0.1070 sq.mm. with a penetrator diameter of 0.75 mm. is theoretically equivalent to an angle of indentation of 57 degrees. The diamond pyramid has an equivalent angle of indentation of 38 degrees so that DPH numbers should theoretically be lower than derived Monotron values. This is doubtless part of the explanation of the authors' results in their Table II.

⁸R. G. Batson, *P. I. Mech. Eng.*, Vol. 1, 1923, p. 401.

⁹A. Krupkowski, *Revue de Metallurgie*, Vol. 28, 1931, p. 641.

Oral Discussion

S. L. HOYT:¹⁰ I do not believe I will take any time to congratulate the authors, although they certainly deserve it. They have given us the hardness conversion relationships in a better fashion than we have had them before, and we are indebted to the authors for having taken the time and trouble to prepare such an excellent treatise as this.

I would suggest to the authors that they take into account the somewhat peculiar characteristics of the property that they are dealing with. I refer particularly to the fact that hardness is more than a property; it is a behavior. When you attempt to convert from one scale to another or from a hardness reading to tensile strength, for example, you run into a difficulty due to the fact that an expression of hardness requires two parameters or two constants. The use of a single number may lead to an error at times which may be very confusing. I feel it safe to say that if we were to use these conversion factors for steels which have been cold-worked that we would run into such difficulties and that would be particularly true if we attempted to convert to strength.

The diamond pyramid hardness is credited here with more of a basic characteristic than I believe it deserves. It is true it has an excellent scale, from low loads to high loads, but due to the geometry of the impressions produced at different loads you always measure the hardness or the resistance to penetration at the same degree of cold work, whereas if you were to use a ball indenter you would depart violently from that circumstance and would actually be measuring the resistance to deformation or hardness at different degrees of cold work. It is well known that cold work changes the properties of materials, and hence if we leave out a factor which is as fundamental as that, there is something lacking. I do not mean to say that that takes away from the value of this paper but the restriction which is imposed by that circumstance I think should be brought out. This is all made clear by means of the so-called Meyer analysis of ball indentation hardness. Meyer also studied the cone indentation hardness test of Ludwig. That is very closely what the diamond pyramid and Rockwell hardness tests are so that this factor could very well be taken into account.

There is one other point which I would like to mention in passing, and that is that when dealing with cemented carbides, the geometry of the impression changes at something like 90 on the Rockwell A scale. While the Rockwell brale is called a cone, the end of it is a sphere, and for small indentations you are dealing with a spherical indenter and not a conical indenter. As I recall it, when the Rockwell A hardness number is around 92 or above, the load on the indenter divided by the projected area of the impression (the hardness number) comes of the order of 2200 to 2400 kilograms per square millimeter. That measurement is independent of any of the peculiarities of the Rockwell because the load on the diamond point and the area of the impression were both determined. The mean pressure is a little higher than the figures reported by the authors. Personally I do not see anything wrong with the determination of that mean pressure as a measure of the hardness.

¹⁰Member of Staff, Battelle Memorial Institute, Columbus, Ohio.

H. W. GRAHAM:¹¹ In our laboratories we have done a great deal of work on hardness and impact strength in relation to cold work and it was not done to explore anything that bears particularly upon this paper. But after hearing this paper presented I can see that it does have a bearing. The type of test which we have made involved both impact strength and hardness, usually Rockwell, from a supposedly fully normalized condition up to ordinarily 10 to 15 per cent cold work, and sometimes as high as 100 per cent cold work.

It was plain to be seen that the hardness was changing very rapidly, and particularly very small percentages of the cold work. That is to say that the slope of the work hardening curve around the area of 0.5 to 2.25 per cent of cold work was quite steep. This fact may be quite important in correlation of different hardness testing methods that vary in the amount of cold working involved.

Authors' Reply

Hardness conversion charts are usually offered with the apology that they are only approximate relations. If the instruments involved are dependable there is no reason apparent now why they should not give accurate and reproducible results within reasonable limitations. That is our position and we are very glad to find material support for it in the discussions of this paper. One glaring exception is recorded, however, by Mr. Kommel, and we are very grateful to him for presenting it. The conclusions to be drawn from his observations are obvious and require no further comment here.

Mr. O'Neill reports excellent agreement with the British Standards Institute data except at very high values. Mr. Gray notes that the divergence between our data and that of the Wilson Mechanical Instrument Company does not exceed 1.5 per cent. Both Messrs. Spring and Gray itemize essential precautions which make such agreement possible. Mr. O'Neill gives an enlightening analysis of the standard hardness tests and a quite acceptable explanation of why the hardness number calculated from Monotron hardness readings is high relative to the observed DPH. Unfortunately, none of these commentators brings up a sharp issue from which an illuminating argument can be developed.

Both Dr. Hoyt and Mr. Graham raise questions as to the effects of cold working on conversion relations. This has been recognized as a distinct problem which is even more important in the case of nonferrous alloys than in that of steel. Another similar question is involved in the comparison of steels having the same tensile strength, but differing markedly in yield point. We hope to investigate both problems eventually.

Dr. Hoyt brings up an issue which appears to be chiefly one of viewpoint. His viewpoint is broad; our concept of hardness is narrow, intentionally so for the purpose of avoiding academic discussions which are beyond the scope of our efforts. Certainly one should attempt to learn as much as possible from a single test, but, as Dr. Hoyt points out, measurement of another parameter is necessary to evaluate the work hardening characteristic. The Knoop-Peters¹² indenter has interesting possibilities in this direction. It should be emphasized,

¹¹Director of metallurgy and research, Jones & Laughlin Steel Corp., Pittsburgh.

¹²National Bureau of Standards Research Paper RP1220, July, 1939.

however, that any such complication of the hardness test is hardly possible in the shop. Other discussions have brought out the detailed requirements necessary to get reproducible results from measurement of one parameter, and the difficulty of assuring that they are observed and familiar to all. We readily agree with Dr. Hoyt that the hardness number of a cemented carbide determined from the diameter of a Rockwell "A" scale impression is an acceptable measure of hardness, but are sure that if he insists on using this measurement, someone will eventually want it converted to some other scale.

The real test for a hardness conversion table is whether or not other observers can check it. One example of agreement is given in the text, and another noteworthy one has come to our attention since preparing this paper. This recent contribution is published by the British Standards Institute under the title of "Approximate Comparison of Hardness Scales," and was obtained through the courtesy of Mr. Hugh O'Neill.

A comparison of these recent observations for Rc and DPH are given in Fig. 9. There is very good agreement, within 1.5 per cent for hardnesses up to 1000 DPH, 69 Rc, between the Wilson chart No. 38 and the Westinghouse relationship. The agreement between the B.S.I. data and the Westinghouse and Wilson data is not so close. The discrepancy may be due to the averaging of data taken from nonuniform specimens as the tables give variations of 4 Rc for a particular DPH. A footnote is added in the British tables stating that these variations are to be expected among individual cases. A difference of 4 Rc is certainly too much if an accurate conversion relationship is desired. If the maximum values of Rockwell hardnesses given in the tables are plotted, as shown in Fig. 9, they agree very nicely with the Westinghouse and Wilson values. The more common errors in testing indicate that maximum values are more nearly correct than the average.

The American Society for Metals conversion relationship shows the greatest disagreement of the compared data. This relationship presumably originated from the conversion table in the 1935 edition of the SAE handbook. The discrepancy is large, amounting to 3 Rc for hardened steels and probably arises from referring all other hardness test values to an unstandardized Brinell scale.

SOME PRACTICAL NOTES ON HIGH SPEED STEEL FORGINGS

By W. H. WILLS

Abstract

After stating that the 18-4-1 type of high speed steel predominates as a material for forgings as well as bar stock used for cutting tools, the advantages of forgings over bar stock are pointed out.

Due to the dense nature of the alloy and the wide range of sizes in which it is required, some difficult mill problems arise, especially when larger sections are involved.

The principal points concerned in the manufacture of high speed steel upset forgings are described. Comment is made on forgings as compared to bar stock and the results of tests comparing impact values of forgings with bar stock of a corresponding size.

HIGH speed steel is one of the principal materials from which metal cutting tools are made. For years the 18-4-1 type has predominated as the most popular type of high speed steel for general cutting tool purposes although more recently there has been some replacement by molybdenum high speed steels, and carbide materials. With the constantly increasing efficiency expected of such tools, their manufacturers must have a high standard of quality in the high speed steel which is their raw material. The chemical and physical requirements, especially that phase of the latter dealing with structure, have become more important than what was formerly required and call for the best of care on the part of the mill if complaints and rejections are to be avoided.

High speed steel is furnished in the form of hot-rolled or hammered annealed bar stock, cold-finished bars, and forgings. The first division includes as common sizes $\frac{1}{4}$ inch up to 8 inch round or equivalent, although larger sizes have been produced. Sizes up to 4 or 4.5 inch round are generally rolled and those larger than this are hammered. As the size increases, the problem of uniformity

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of carbide distribution becomes more difficult. The use of forgings in place of the larger sizes of bar stock is now quite general among the cutting tool manufacturers, although much depends on the type of tool and the length or thickness with respect to the diameter. The advantage of forgings over bar stock lies in the fact that a more homogeneous structure is possible, resulting in finished tools of greater strength and cutting quality.

It is the purpose of this paper to cover some of the main points in the process of manufacture of high speed steel upset forgings that are used for various kinds of cutting tools and discuss some of the problems that arise in connection with the production of these forgings. A brief review of the metallography of high speed steel will be helpful for a better understanding of what is involved in the production of quality high speed steel forgings.

In general the structure of high speed steel differs from carbon and low alloy steels in the size and character of the carbide segregate. The structure of annealed high speed steel is composed of a matrix of ferrite through which is dispersed particles of carbide segregate. These particles are of two general sizes—the larger representing primary carbide persisting from the original cast structure, and the smaller ones resulting from reprecipitation from the solid solution each time the steel has been cooled following heating above the critical points. The primary carbide exists in the original ingot in cellular or network form. J. P. Gill, in his Campbell Lecture dealing with the Carbide Segregate of High Speed Steel, mentions that factors affecting the character and distribution of this segregate include ingot size, mold design, casting temperature, rate of cooling, and degree of mechanical work. The coarseness of the carbide network is much affected by the rate of cooling from the liquid state. The comparatively rapid rate of cooling in small ingots produces a fine carbide network, and the slower rate which prevails as the mass of metal increases results in a coarse network with larger ingot size (Fig. 9). As the ingot size increases, the interior structure is considerably coarser than the exterior. In contrast with the cementite of carbon and low alloy tool steels, which readily goes into solution when heated above the critical range, the carbide segregate of high speed steel is quite refractory, and the larger particles do not completely dissolve even when heated to the hardening range. The only way the carbide segregate can be broken up and redistributed is by hot working such as forging or rolling.

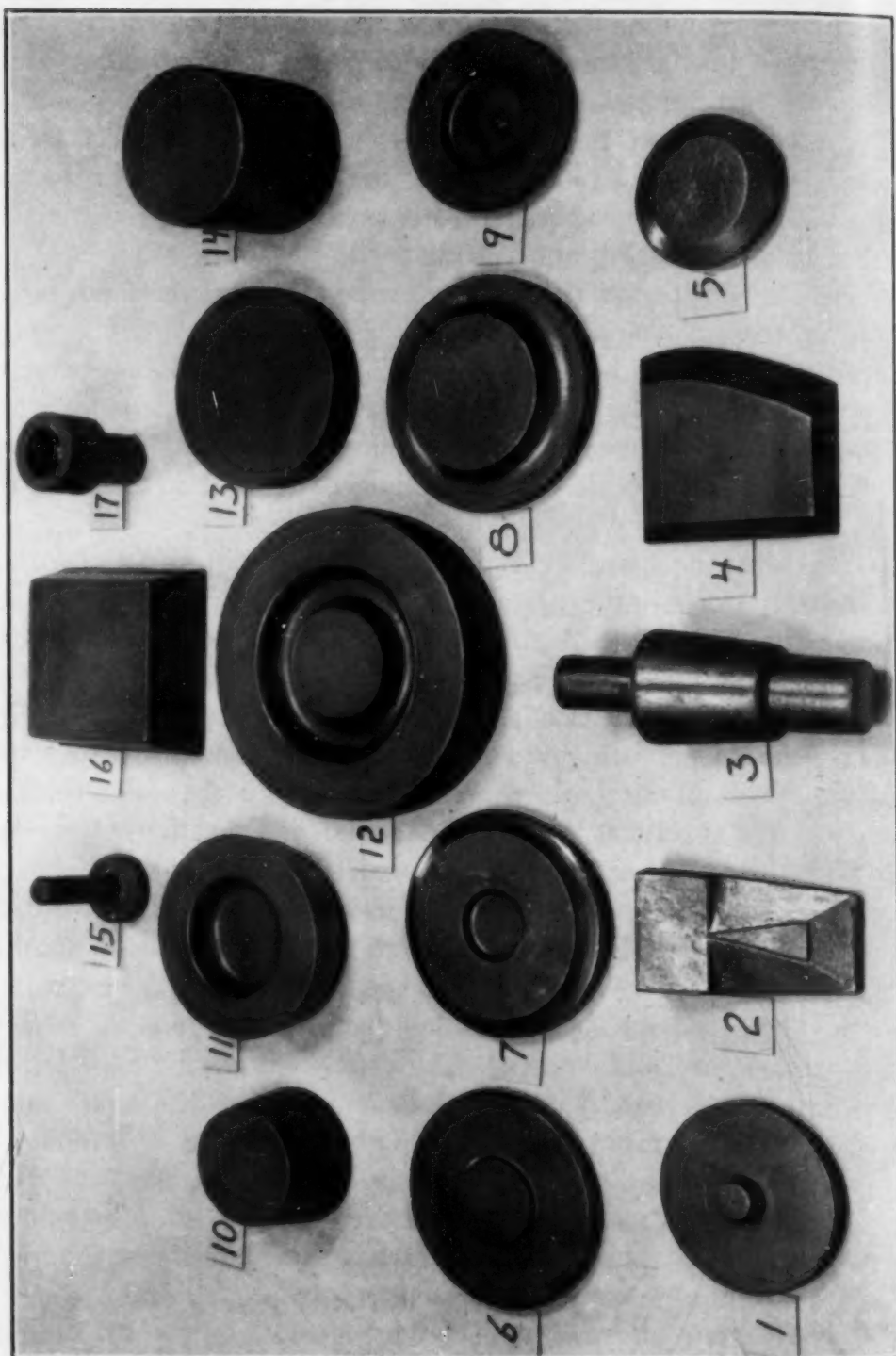


Fig. 1—Group of Typical High Speed Steel Forgings from which the following are made: 1. Meat Cutter Plate, 2. Sugar Cane Shredder Knife, 3. Plunger for Molding Glass Insulator, 4. Wood Heel Knife, 5. Involute Gear Milling Cutter, 6. Slitting Cutter, 7. Involute Gear Milling Cutter, 8. Blanking Die, 9. Gear Shaper Cutter, 10. Tire Turning Tool Insert, 11. Gear Shaper Cutter, 12. Recessed Milling Cutter, 13. Plain "Pancake" Forged Disk, 14. Hob, 15. Woodruff Keyway Cutter, 16. Tire Turning Tool, 17. Gear Tooth Chamfering Cutter.

High speed steel is cast in molds of various sizes—those in common use including:

Size—Inches	Approximate Ingot Weight Pounds
6	300
8	750
12	1600
16	3300

big end up with more or less of a taper. The ingot structure as cast is quite brittle although when worked down into billets the metal exhibits more ductility. The ingots are carefully heated and the total



Fig. 2—Bar Blank, Forging and Finished Tools. No. 1. Shell End Mill, No. 2. Woodruff Keyway Cutter.

heating time runs about 4 to 1 compared to carbon tool steel heating practice. For billets this ratio is approximately 2 to 2.5 to 1.

As stated above, the carbide segregate is broken up in the hammering and rolling operations. In the hammering of ingots down into billets (known as cogging) and the hammering or rolling of billets into bars, the metal is elongated with little or no transverse working. Consequently the cells of the carbide network are elongated and carbide particles have more or less of a tendency to be arranged in lines. In general the smaller the size, the more uniform the carbide distribution, and in sizes such as 1-inch round or less, comparatively few carbide particles are distributed in lines.

On account of the dense nature of high speed steel, the reduction per pass on hammer or mill is light as compared to soft steel practice. This, together with long heating cycles and structural considerations, limit the size of the ingot, and the larger sections of necessity get relatively less working from ingot to finished bar than in the case of soft steels. The effect of the hammer blow does not

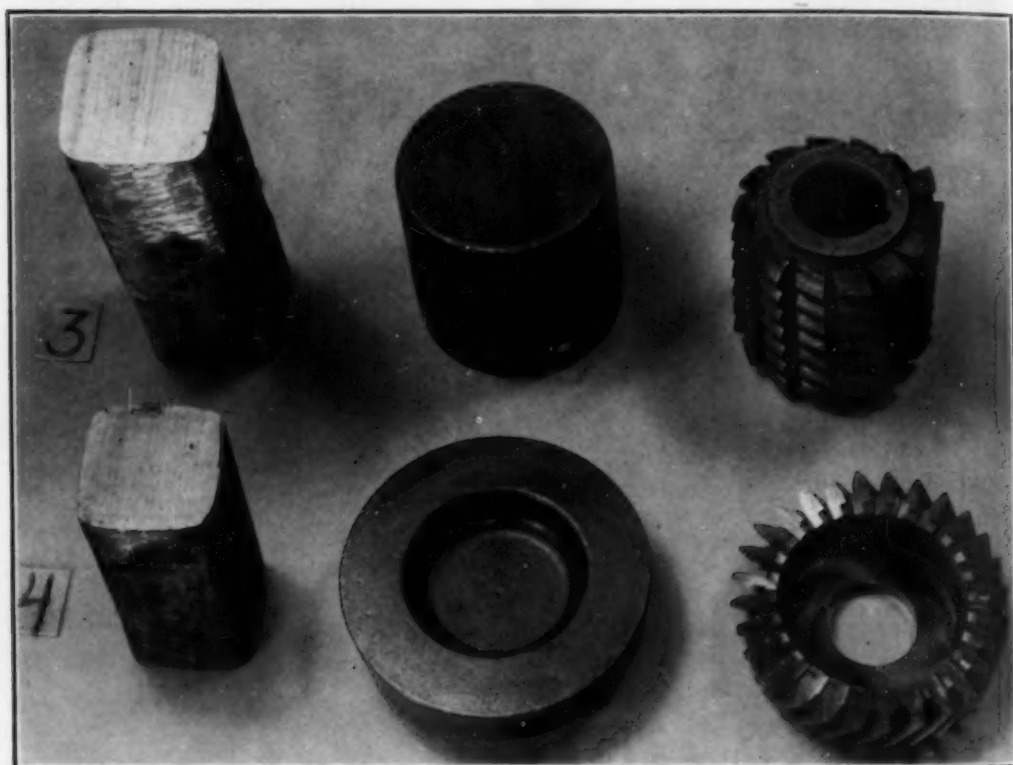


Fig. 3—Billet Blank, Forging and Finished Tools. No. 3. 3x3-inch Hob, No. 4. Spiral Gear Shaper Cutter.

penetrate a large high speed steel section like it does soft steel, so there is more difference in grain refinement from the exterior to the center. This means that such large sizes are apt to show a center structure with a more or less cellular distribution of the carbides.

From this, one can appreciate the mill's problem in obtaining satisfactory carbide distribution over this wide range of bar stock sizes. Also the advantage of upset forgings can readily be seen—

- (1) The metal is worked transversely as well as longitudinally, resulting in more uniform carbide distribution.
- (2) By starting from a relatively small billet from a smaller ingot than would be used for bar stock of a given size—primary carbide particles are smaller.

These are factors of prime importance in the production of cutters of maximum strength and cutting quality.

Most high speed steel forgings are made from the 18-4-1 type, to a lesser extent 18-4-2 and the molybdenum types, and still less from cobalt high speed steels. The forgeability of the 18-4-1 and some of the molybdenum types is better than the others, that is, they will withstand upsetting and rolling operations better without bursting.



Fig. 4—Billet Blank, Forging and Finished Gear Tooth Cutter.

For the most part, high speed steel forgings are worked from round cornered billets and to a lesser extent from round bar stock. Sizes commonly used range from 1.5 up to 6 inches square. The annealed billets are carefully inspected for internal defects by deep etching and examination of the fracture of hardened disks. The structure is checked by microexamination of longitudinal sections (Fig. 10). A sound center condition is necessary to withstand the severe upsetting operation. Also billets should have a smooth surface free from corner checks and defects which require deep grinding for their removal. The latter is especially objectionable as the weight of the stock for a cutter blank must be accurately figured, and such a condition may be a source of error in weight.

After the billet stock is cut to proper length, the pieces are preheated at about 1400 degrees Fahr. and carefully brought up to the forging heat (range 2050 to 2100 degrees Fahr.). They are shifted

about to insure uniform heating. The time at the forging heat must be watched carefully so as to avoid excessive decarburization. The usual size hammers used range from 1500 to 5000 pounds and the stock is upset on flat dies. The finishing may be on flat dies or shape dies depending on the kind of forging. Where the shape is such that a slight flash is formed, it is removed by hot trimming.

After finishing, the forgings are allowed to cool slowly in sil-o-cel. They are then box annealed at 1600 to 1650 degrees



Fig. 5—Billet Blank, Forging and Finished Flange and Tread Finishing Tool.

Fahr., holding an average of 8 to 10 hours at temperature, and cooled slowly in the furnace. Owing to the extremely dense nature of the metal in the forgings, the hardness after annealing averages slightly higher than fully annealed bar stock.

For convenience, upset forgings may be divided into two classes, depending on how they are finished.

- (1) Those finished with flat dies. (Class 1)
- (2) Those finished with shape dies. (Class 2)

Class 1 includes all plain flat upset disks with thicknesses less than one-half the diameter. Sizes over 5 inches in diameter are not made in large numbers, so the cost of a die is hardly justified. Di-

ameters as large as 10 inches or more are required occasionally but not often. The smaller sizes are often called "pancake forgings." There is really no advantage in using a shape die on these if they are finished properly on the edges. They are upset from a length of billet that must be at least twice the billet size. The rounded edges are then squared and the exterior structure further worked by the so-called rolling operation. The rolling may be interrupted by a few blows on the flat surface as the disk approaches finished size.



Fig. 6—3.5-Inch Diameter Forging Made With Flat Dies Showing Internal Hammer Burst. Such Defects may originate from heavy Carbide Segregation, Improper Heating, or Working at Too Low a Temperature.

Class 2—The use of shape dies for the finish forging operation is applied to quite a range of cylindrically shaped forgings as well as to special shapes. By means of these dies, the size can be held closer and a smoother finish obtained. Furthermore, there are some cases where by their use there is less danger of checks on the exterior, or of developing internal bursts, as the pressure can be more uniformly applied.

When shape dies are employed, the stock is upset on flat dies and rolled on edge roughly to size. This may be repeated one or more times, but with multiple upsetting, the chance of hammer bursts in the center increases. The forging is finally compressed in the die. Where it is possible the forging should be rotated and given several blows in this operation. The metal adjacent to the

outer circumference is squeezed into close contact with the die and thoroughly compressed at the vital zone, which later becomes the cutter teeth.

As previously stated, due to the dense nature of the metal, the thickness affected by the hammer blows is much less than with soft steel. Consequently as the thickness and diameter of the disk increases, the structure in the center is less affected in the upsetting operation and has a tendency to retain more or less of its original

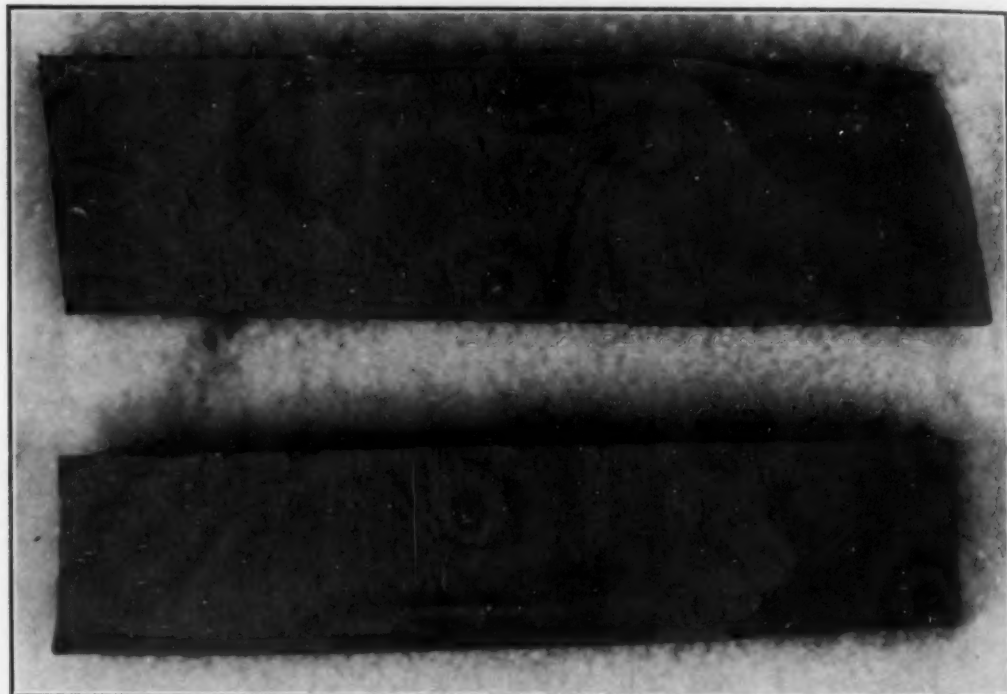


Fig. 7—Hardened Fractures of 2.5 inch square Billets Showing Rather Heavy Carbide Segregation.

brittle nature. It is a matter of experience that when the thickness of a forging exceeds one-half the diameter there is an increase in the chances of developing hammer bursts during the rolling operation on the flat dies. When such forgings are finished in shape dies the pressure is applied uniformly and danger of hammer bursts is largely eliminated. For this reason cylindrically shaped forgings with lengths 50 to 100 per cent of the diameter are preferably finished in dies. Lengths required do not often exceed the diameter. Such forgings are the most difficult to make—that is, to obtain uniform carbide distribution throughout and sound center structure. Common diameters handled in shape dies run from 2.5 to 5 inches in $\frac{1}{4}$ -inch steps. The popular hob cutter sizes are 3 to 4 inches inclusive.

COMPARISON OF FORGINGS WITH BAR STOCK

Having discussed the principal points regarding the production of high speed steel forgings, it may be in order to offer some comments in the way of comparison between forgings and bar stock as a cutting tool material. That the manufacturers recognize the advantages of forgings for milling cutters, hobs, and many other kinds of cutting tools is indicated by their widespread use. There is

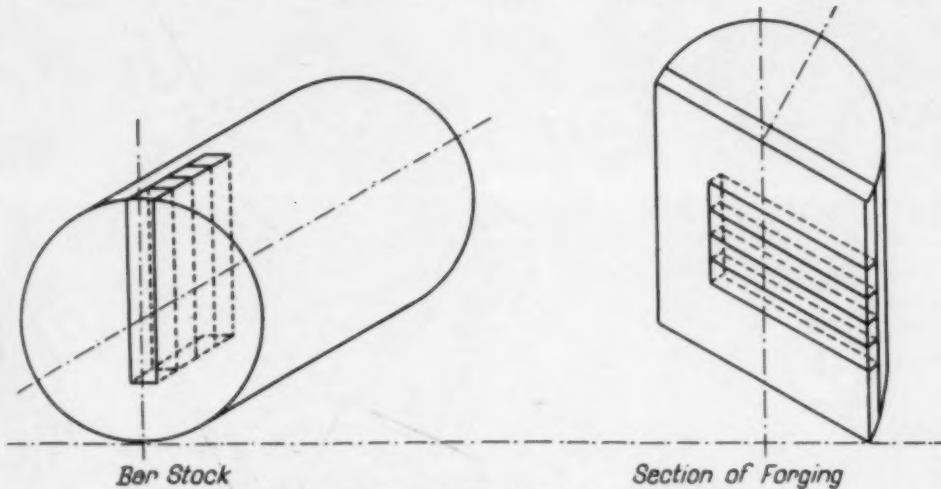


Fig. 8—Sketch Showing Method of Taking Impact Tests.

some difference of opinion as to the limiting size where the change should be made from bar stock to forgings. It is doubtful whether any decided improvement in structure is gained under 3-inch round but as the diameter increases to 3.5 or 4 inches the advantage is more pronounced. This continues as the diameter goes up and the thickness is less than 50 per cent of the diameter. Sizes of 7 inches in diameter and over and with lengths over 50 per cent of the diameter, any advantage over bar stock is doubtful. These statements are based on a study of microspecimens taken from a number of forgings ranging from 3 inches in diameter by 1.5 inches thick and 3 inches in diameter by 3 inches thick up to 7 inches in diameter by 3.5 inches thick and 7 inches in diameter by 5 inches thick (Figs. 11 and 12).

In connection with this investigation of the structure of forgings, a series of impact tests were run with the idea of comparing bar stock with forgings of the same diameter. These were all of 18-4-1 of similar analysis. Four unnotched Izod impacts test were machined from the bar stock and forgings of each diameter, as per the accompanying sketch (Fig. 8). In order that the 120 foot-

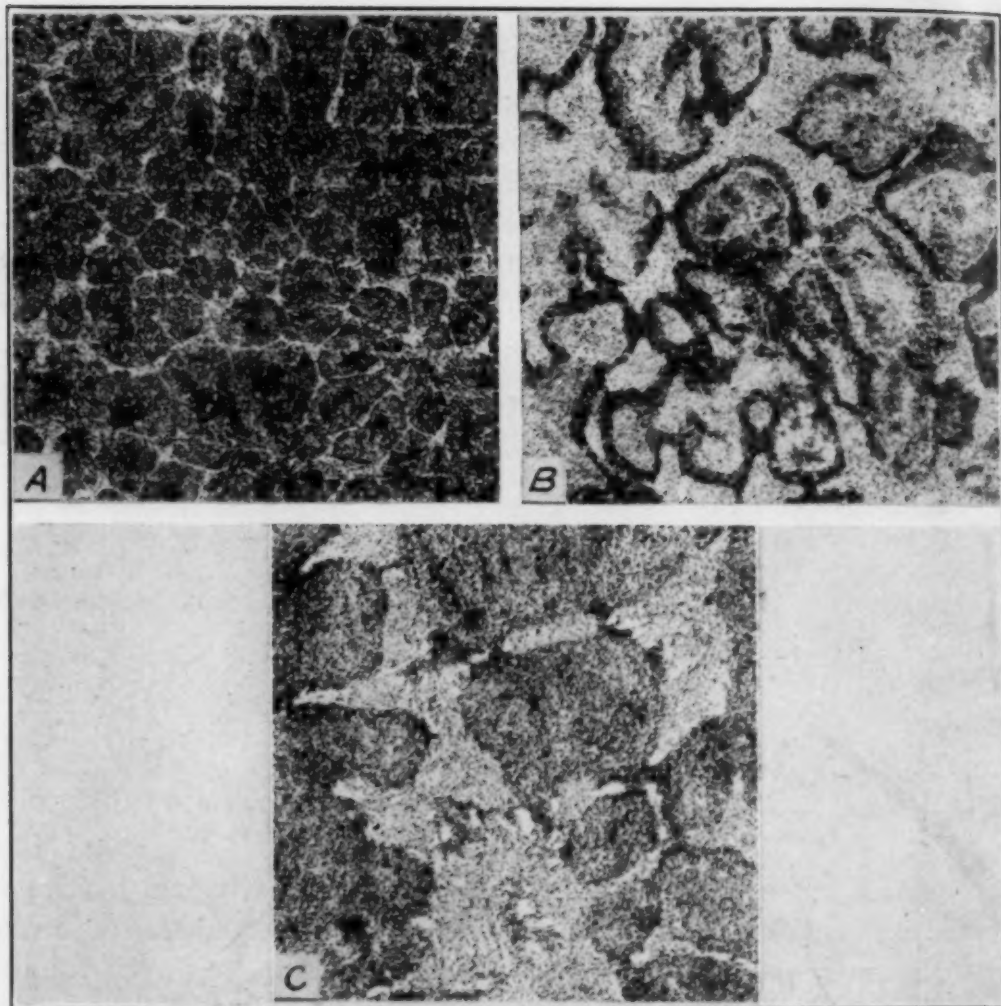


Fig. 9—High Speed Steel (18:4:1) Ingot Structures Annealed. $\times 125$. Fig. 9A—Section From 3-Inch Square Ingot Showing Typical Fine Cellular Carbide Distribution. Fig. 9B—Section From 8-Inch Square Ingot Near Exterior Showing Typical Cellular Carbide Distribution. Fig. 9C—Section From 8-Inch Square Ingot Near Interior at Pipe End Showing Relatively Coarse Cellular Carbide Distribution.

pound blow might be applied at a point that would represent the base of a tooth, the specimens were specially clamped in the machine so that this point was $\frac{3}{4}$ inch in from the exterior of the bar stock or forging. The blow was struck tangentially to the axis in each case, as indicated in the sketch. It is realized that the foot-pounds of energy absorbed is not a true impact value but as all tests were run in the same way, a comparison could be obtained. All impact tests of specimens were given the same heat treatment in a gas-fired controlled atmosphere furnace, 4 test pieces handled at a time. This was as follows: Preheated at 1600 degrees Fahr. for 10 minutes, transferred to the high heat at 2325 degrees Fahr. for 2 min-

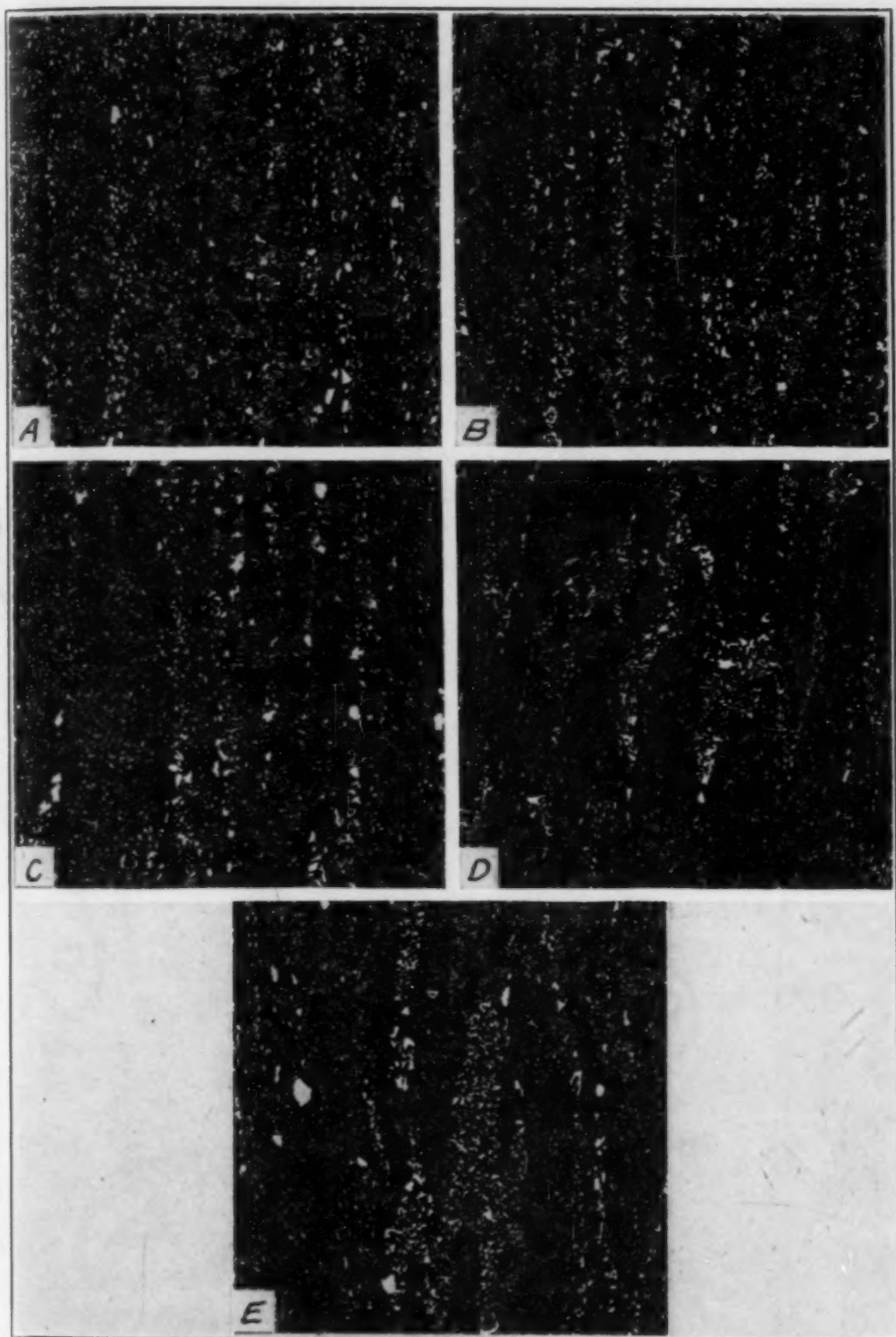


Fig. 10—High Speed Steel (18-4-1) Billet Structures. $\times 125$. From Various Forging Sizes, Showing Typical Carbide Distribution About Half Way Between Center and Surface. There is More or Less Tendency for the Carbides to be in Lines Parallel to the Axis of the Billet. A—1 $\frac{1}{4}$ -Inch Square; B—2 $\frac{1}{2}$ -Inch Square; C—3-Inch Square; D—4-Inch Square; E—5-Inch Square.

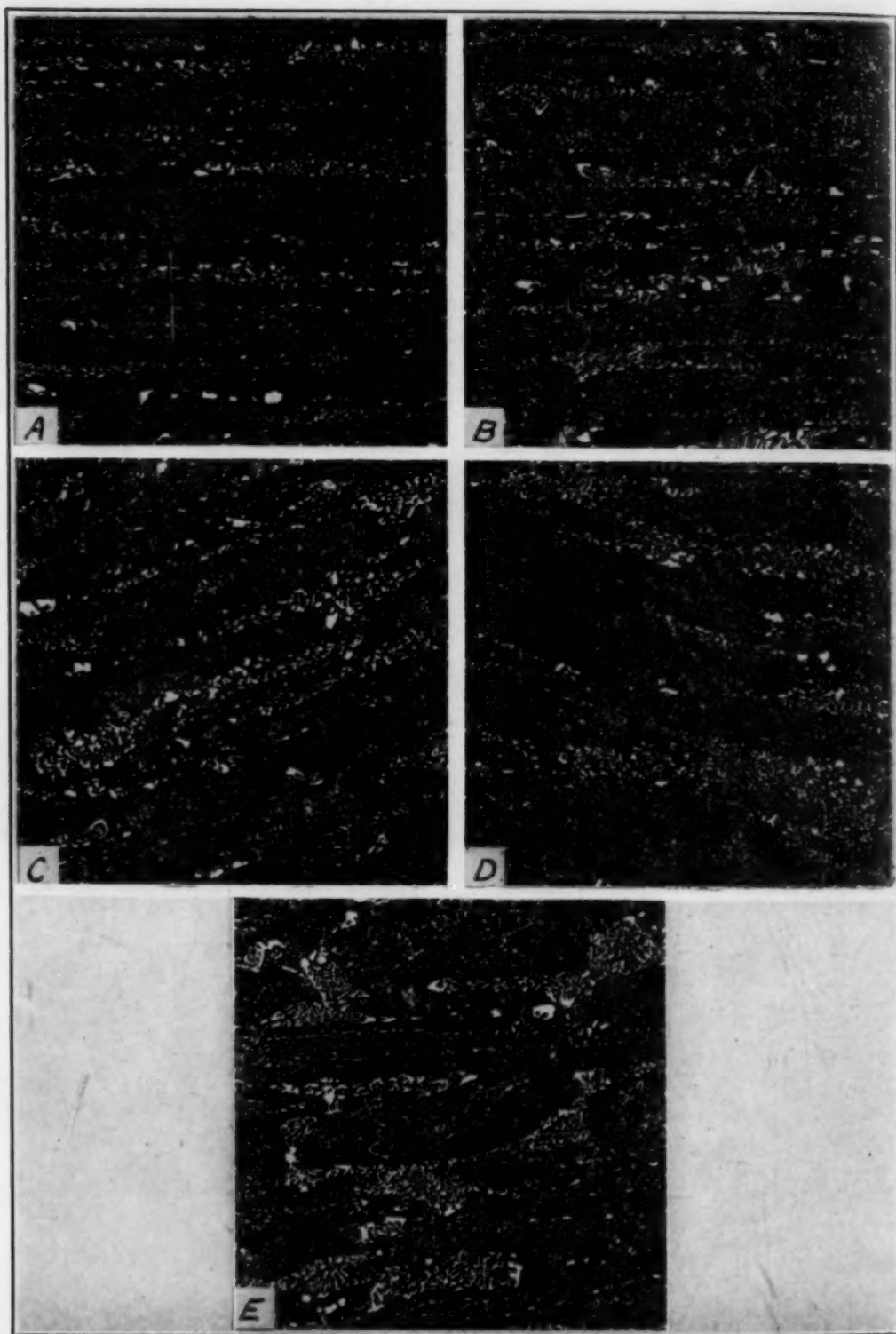


Fig. 11—High Speed Steel (18-4-1) Round Bar Structures. $\times 125$. From Various Sizes Showing Typical Carbide Distribution About $\frac{3}{4}$ Inch in from the Surface. As in the Case of Billets, the Carbides are More or Less Distributed in Lines Parallel to the Axis of the Bar. A—3-Inch Round; B—4-Inch Round; C—5-Inch Round; D—6-Inch Round; E—7-Inch Round.

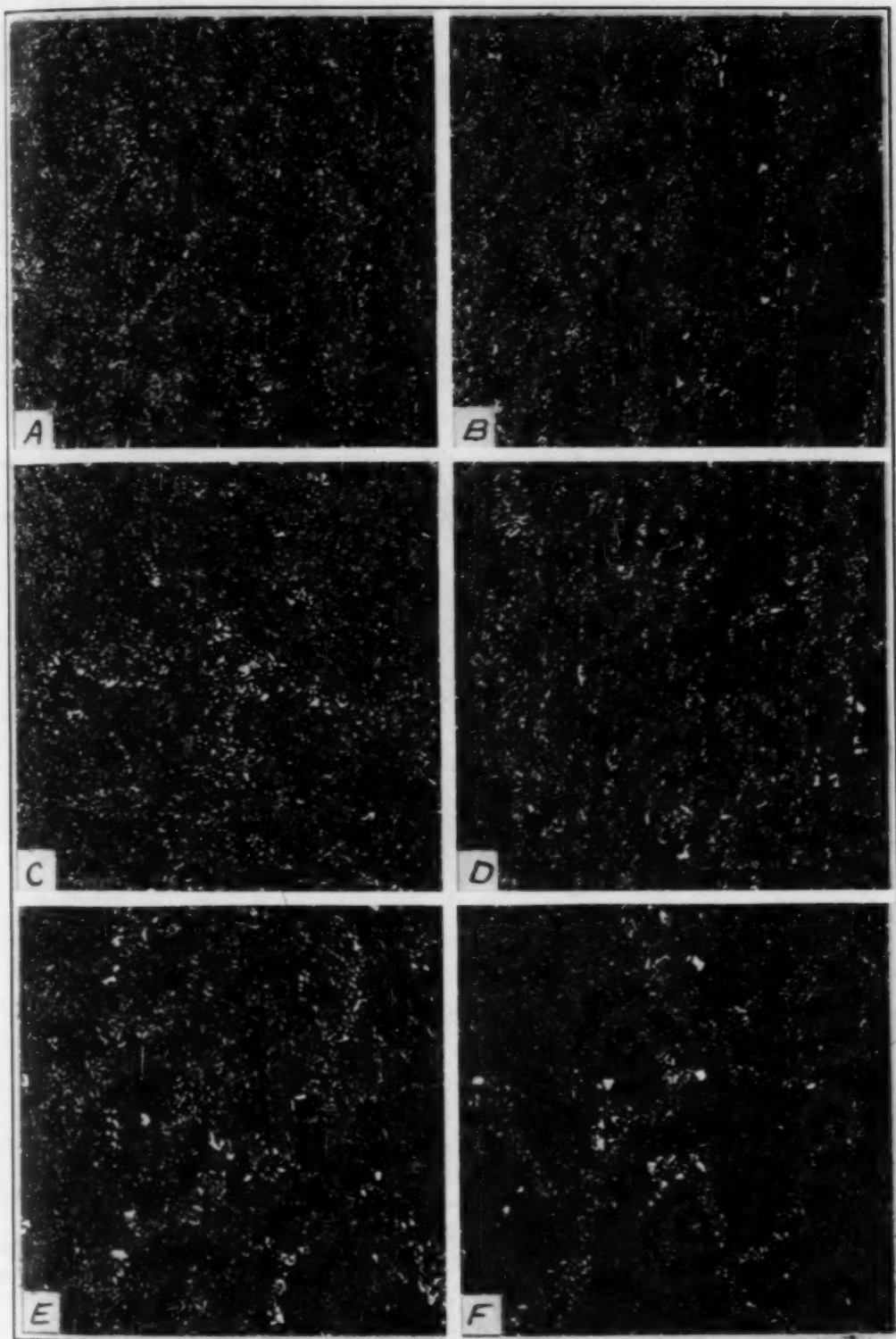


Fig. 12—High Speed Steel (18-4-1) Forgings. Typical Structure of Upset Forgings of Various Sizes Showing Distribution of Carbides. Microspecimens Were Taken From Longitudinal Sections Through the Axis of the Forgings About $\frac{3}{4}$ Inch in from the Exterior and Half Way Between the Ends. $\times 125$. As Compared to Bar Stock There is More Uniform Carbide Distribution and Average Size of Primary Carbide Particles is Smaller. A—3 Inches in Diameter $\times 1\frac{1}{4}$ Inches Thick; B—3 Inches in Diameter $\times 3$ Inches Thick; C—4 Inches in Diameter $\times 2$ Inches Thick; D—4 Inches in Diameter $\times 4$ Inches Thick; E—5 Inches in Diameter $\times 4\frac{1}{2}$ Inches Thick; F—5 $\frac{1}{2}$ Inches in Diameter $\times 2\frac{1}{4}$ Inches Thick.

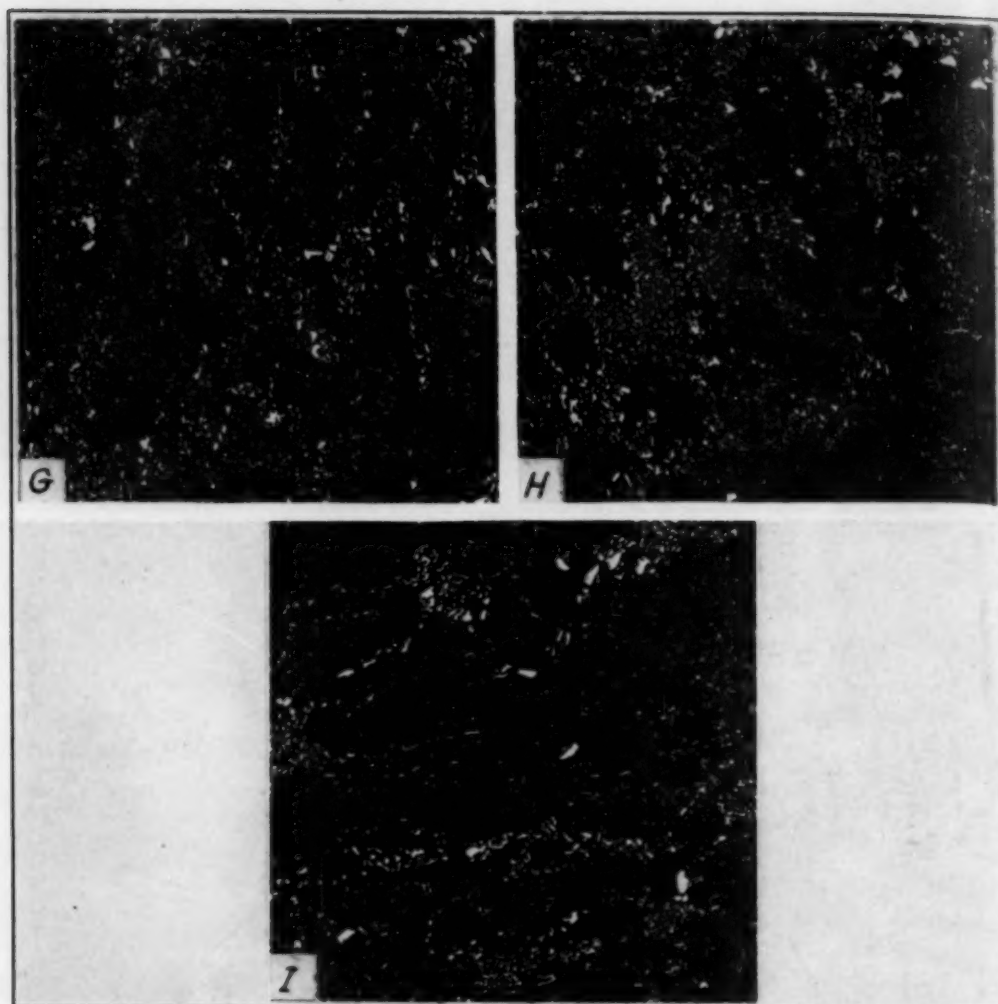


Fig. 12 Cont.—G—6 Inches in Diameter \times 4 Inches Thick; H—7 Inches in Diameter \times $3\frac{1}{2}$ Inches Thick; I—7 Inches in Diameter \times 5 Inches Thick.

utes 15 seconds. After quenching in oil, they were drawn at 1050 degrees Fahr. for 2 hours. These were checked for hardness and run on the Izod machine. Results shown in Table I.

Table I

Size	Bar Stock		Size	Forgings	
	Rockwell C	Average Ft. Lb. Absorbed		Rockwell C	Average Ft. Lb. Absorbed
3" rd.	63.5	9.0	3 \times 1 $\frac{1}{2}$	64.0	7.7
4	64.0	5.4	3 \times 3	64.0	5.5
5	64.0	5.0	4 \times 2	63.8	8.7
6	64.0	5.0	4 \times 4	64.1	4.6
7 $\frac{1}{8}$	64.0	5.0	5 \times 2 $\frac{1}{2}$	64.0	6.6
			5 \times 5	63.8	6.2
			6 \times 4	64.0	5.6
			7 \times 3 $\frac{1}{2}$	64.0	4.7
			7 \times 5	64.0	5.9

The results of this test show that as a whole the forgings have better impact values than the bar stock. This may be partly due to the fact that in these cylindrical forgings the direction of the grain changes from the center to the outside, the exterior resembling that of longitudinal bar sections and the interior that of a bar cross section. This effect would vary, depending on the thickness of the forging, and probably accounts for the variation of the forging impact values whereas those of the bar stock are quite uniform.

It is hoped that the foregoing remarks will give some idea of the essentials involved in the production of high speed steel forgings. Quality forgings as well as bar stock are the foundation for quality cutting tools, and research aiming toward higher standards is constantly going on. This is helped by co-operation on the part of the mill and the cutting tool manufacturers. For the latter, this is just the first step, to be followed by such important ones as proper tool design, heat treatment and grinding.

While we have in recent years seen the introduction of a number of new high speed steels which have their points of superiority in one way or another, it will probably be some time before the position of 18-4-1 is seriously challenged. This holds true particularly for cutting tool forgings. Under such circumstances, research work for the further improvement of 18-4-1 is amply justified.

ACKNOWLEDGMENTS

The author wishes to express his thanks to F. B. Lounsberry, Vice President of the Allegheny Ludlum Steel Corporation, for permission to publish the information given in this paper. He is also grateful to J. F. Bodmer, foreman of the Forge Department (Dunkirk Plant) for helpful suggestions, and to L. R. Lounsberry, of the Metallurgical Laboratory, for his part in conducting physical tests and metallographic work.

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2. A. F. McFarland, "High Speed Steel Drop Forgings for Milling Cutters and Special Tools," *Journal, American Steel Treating Society*, March 1920.
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DISCUSSION

Written Discussion: By W. R. Frazer, Union Twist Drill Co., Athol, Mass.

Mr. Wills has clearly stated the problems encountered in the manufacture of the larger diameter high speed steel bars and explains the metallurgical reasons for using forgings instead of bar stock for those sizes where improved carbide distribution is attained. In Fig. 11 he shows the carbide distribution in various bar structures and in Fig. 12 he shows the microstructure of various forgings of similar diameters to the bars. It is quite evident that the forgings show greater freedom from axial segregation of the carbides and there is a more uniform distribution of smaller carbides. The Izod values in Table I do not reflect the structural improvement in the forgings as the photomicrographs would indicate. This may be caused by the excess capacity of the Izod machine used (120 foot-pounds) and it would be interesting if Mr. Wills would repeat this test with the use of an impact machine of lower capacity.

Mr. Wills has not stressed the effect of upsetting on the flow lines in the forgings. It has been our observation that the upsetting of the billet will definitely change the flow lines from an axial direction (as in bar stock) to a condition where they are rounded at the corners of the cross section and curve into the ends of the forging. Obviously, the more upsetting the forging receives, the more change of axial flow takes place. In pancake forgings, where the thickness is less than half the diameter, the flow lines will be changed by 90 degrees so they turn at the corner to being parallel to the side of the forging. A pancake forging, therefore, will have flow lines that are parallel to the shape of the cutter tooth. This in itself will strengthen the tooth. Furthermore, the additional hot working will effect better carbide distribution and refinement so that the resulting structure is much more uniform as evidenced by the greater refinement shown by the hardened and tempered fracture.

We agree with Mr. Wills that upset forgings will produce better cutters than bar stock in sizes above 3-inch diameter and where the length does not exceed the diameter.

Written Discussion: By Victor Stefanides, metallurgist, Illinois Tool Works and Shakeproof Lock Washer Co., Chicago.

The designer of high speed steel tools that will have to withstand maximum stress in service as well as the forger concerned with grain flow, refinement, and distribution of the primary carbides, should consider this paper carefully.

On reading through this paper, however, one can readily see that no special emphasis is being placed on upset forgings finished with "shape dies." Indeed, the author even commits himself on the top of page 431, to the following:

"There is really no advantage in using a shape die on these if they are finished properly on the edges."

There will be a vast difference between the quality of forgings finished in "shape dies" and those finished with flat dies. In "shape dies" the metal will be compressed, have denser structure, more uniform hardenable characteristics, better machinability, higher physical properties, and be subject to less

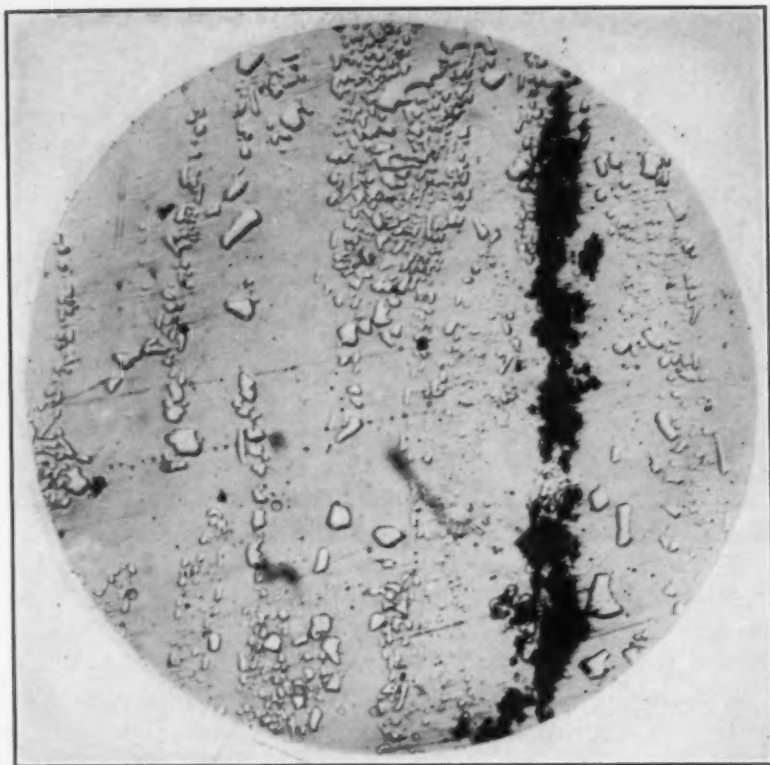


Fig. 1—Photomicrograph Showing Nonmetallic Inclusions. $\times 400$.

deformation on the heat treating operation than identical forgings finished with flat dies. Suppose we analyze these qualities in order as listed above.

The metal will be compressed because the external forces applied are in the direction toward the center. The metal is not subjected to plastic flow in lateral direction and due to this condition the voids and cavities will have no opportunity to open up but instead form adhesive bonds. In cases where gas or nonmetallic inclusions, as illustrated by Fig. 1 of this discussion, exist, in forging with flat dies, this may open up as shown in Fig. 2, page 442.

Fig. 2 is a photomicrograph of a cross section of a milling cutter which was upset from a billet blank and finished with flat dies. It was then annealed, machined and heat treated. In the heat treating operation either the external gases diffused inwardly or the dissolved gases in the steel migrated to the void and due to either the expansion or chemical reaction of the gases and the resultant volume increase, actually raised the layer of metal in the form of a blister.

Naturally steel forged in "shaped dies" will result in denser structure whose specific gravity will be of higher magnitude than that forged with flat dies. The reason for this is again the formation of closer adhesion of the voids, cavities, and compression.

The writer believes that more uniform hardenability of metal forged with shape dies is mainly due to compression, which results in forcing out some of the gases dissolved in the metal. It is believed by the writer that some

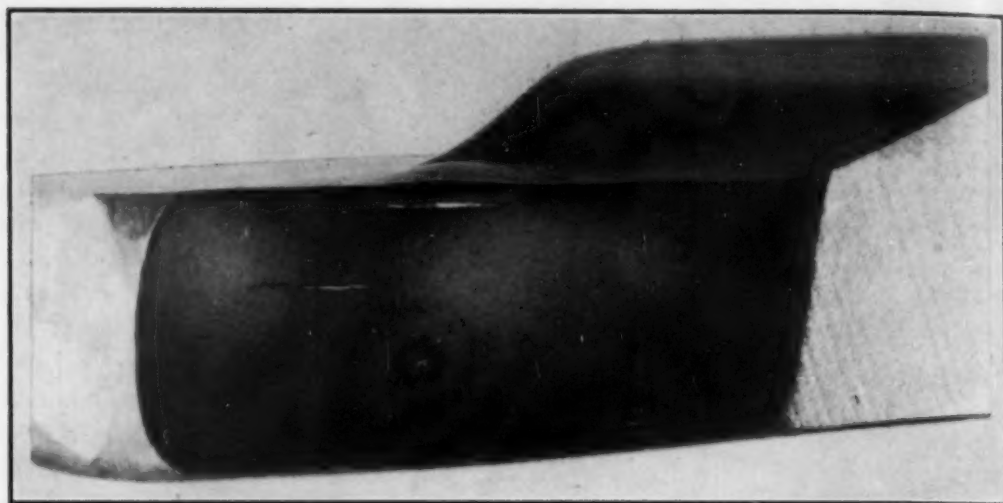


Fig. 2—Photomicrograph of Milling Cutter Blank Showing Surface Cracks. $\times 2$.

gases, such as nitrogen, will hinder the carbides from going into solution, or forcing them prematurely out of solution, while hydrogen will react with carbon to form methane, thus depriving the metal locally of its carbon.

Better machinability is accomplished by reason of compression and adhesion of voids and cavities due to forging with shape dies. The reason for this will be obvious when consideration is given to the fact that the cutting tool has a double duty to perform, namely, closing the cavities with adjacent metal so that it has proper backing and shearing off the metal.

When metal is deprived at least of a small portion of the dissolved gases in compression, the voids and cavities are forced to form stronger adhesive bonds than is possible in forging with flat dies. Having eliminated at least a small percentage of the dissolved gases by compression, their influence in the subsequent heat treating operation will be less effective, resulting in higher ultimate physical properties. Though no specific proof can be cited, there is some evidence that the grain growth is retarded by forging with shape dies, which will also have some bearing on physical properties.

Perhaps the most interesting phenomenon directly traceable to the forging in shape dies or drop forging practice is the fact that distortion can be reduced to a minimum or the change in volume can be predicted. The change in volume usually takes place in the direction of mass; however, distortion on tools made of bar stock may go in any direction with wide variations in volume change. The distortion and the volume change on tools made of forgings with flat dies can be controlled with higher accuracy; however, to accurately predict directional volume change for a given high speed steel and heat treating temperature, it is necessary to resort to forging with shape dies. Here again the reason for this phenomenon may be found in denser forgings, at least partial elimination of gases, and more adhesive bonds of voids and cavities.

Written Discussion: By W. E. Bancroft, metallurgist, Pratt & Whitney Div., Niles Bement & Pond Co., Hartford, Conn.

Mr. Wills' excellent paper furnishes some very gratifying confirmation of

opinions which we have held for some time regarding the superiority of upset forgings over bar stock for certain types of high speed steel cutting tools.

We believe the important feature of high speed steel forgings is the fact that, as Mr. Wills' photographs show, a much more uniform distribution of small carbides and segregate particles can be obtained than is possible in equivalent sizes of bar stock. It seems probable that small well distributed particles of carbides will go into solution easier and quicker during the hardening operation, causing a richer and denser martensitic matrix when the tool has been quenched and tempered. This condition results in a finer grained and denser cutting edge which will stand up in service longer than one wherein the carbides are in the form of large globules possibly arranged in the form of networks or streaks.

Figs. 11 and 12 of this paper give an excellent illustration of the superior carbide distribution and particle size which can be obtained in forgings over equivalent diameters of bar stock. We believe two factors come into play here; first a certain diameter forging can be made from a smaller sized ingot than can the equivalent size bar stock, thus having smaller and better distributed network of segregate to start with, and second the extra hot work in two directions on the forging tends to break up the parallel streaks and network in the billet so as to eliminate planes of weakness.

Mr. Wills' impact figures are interesting. While we do not feel that impact values on hardened tool steel in general are very reliable, we believe that in this case, since a number of readings were made and average results reported, the conclusion that forgings have slightly greater impact strength at the location tested is accurate confirmation of what might be expected from a study of the microstructure.

I do not know whether the various theoretical explanations which have been developed for the superiority of forgings over bar stock offer a complete explanation for the results obtained or not, but I do know that forging small individual pieces of steel seems to impart some desirable property to that part which cannot be obtained in any other way. Many times in our shop we have listened to old-time mechanics who insist that no tool such as chisels, scrapers, punches, tool bits, etc., is any good unless it has been heated up and hammered a little. Probably some of this feeling is prejudice and probably some of it arises from the fact that in past years bar stock quality was not as well controlled as it is today. I do believe, however, that there is something to it and that the same thing applies to high speed steel forgings. Certainly our experience has shown that over a period of time greater average satisfaction can be obtained in service from high speed steel cutting tools made from forgings as compared with bar stock.

Oral Discussion

O. W. BOSTON:¹ I believe it is generally recognized by the manufacturer of cutting tools that those tools made from forged high speed steel give a performance superior to those made from bar stock. I think this is borne out by the rather conventional methods of manufacturing drills, large single point

¹Director, department of metal processing, University of Michigan, Ann Arbor, Mich.

tools, and milling cutters. I have some doubt in my own mind as to the reliability of tool performance, however, when based upon hardness or even the Izod impact value. It occurred to me that the paper might be enlarged and made much more valuable if the author had some data to submit which would show for these three classes of tools, which have different methods of failure, some cutting tool performance. I am wondering if he could substantiate his conclusions based on Rockwell and impact values with some definite cutting performance data.

Author's Reply

Dr. Frazer, in paragraph 2 of his discussion, elaborates on the first mentioned advantage of upset forgings which we agree applies particularly to the so-called "pancake" type of upset forgings, with the diameter relatively large as compared to the thickness.

Mr. Stefanides brings out a number of qualities which are characteristic of die forgings that would not necessarily apply to forgings made in flat dies. Assuming that the proper amount of upsetting and rolling has been done, we do not believe there would be as much difference as is indicated. We find that the major portion of the mechanical working takes place on flat dies and comparatively little metal is displaced in the shape die when finishing. Otherwise an excessive amount of flash would be produced. We believe the principal advantage of the shape die lies in the close size and better finish possible, also there is less chance of hammer burst when handling thick forgings.

Regarding Professor Boston's comments, the general statement might be made that high speed cutting tools made from forgings are less apt to fail by breakage when subject to abnormal strain in service. This applies particularly to multiple point tools such as milling cutters, hobs, and gear cutters, and is based on the experience of a number of cutting tool manufacturers. So far as high speed steel is concerned, Izod impact values are useful only in giving an indication of resistance to shock rather than any tool performance.

HARDENABILITY STUDIES ON TOOL STEEL

By A. J. SCHEID, JR.

Abstract

This paper covers the hardenability characteristics of the principal types of tool steels. Some of the effects of composition changes are pointed out and their usefulness indicated.

The size of the samples chosen and the temperatures used were selected so that the results would be of practical use in the selection and hardening of tool steel.

IN recent years, hardenability of several types of steels has been given attention. The word "hardenability" is used in this paper to mean surface hardening effect and hardness penetration as measured by the indentation hardness method in different sized specimens quenched under given conditions. The relation of composition, grain size, other inherent characteristics and quenching speeds to hardenability are shown in these studies.

When hardenability is discussed in tool steel, the first classification is based on quenching medium, i.e., whether water, oil or air. In the water hardening series (requiring relatively high quenching speed) by reason of class characteristics, the interest is in the mass effect on depth of hardness, core condition and fracture grain size when quenched from a given temperature.

In oil hardening steels (requiring relatively low quenching speed), the interest is in the mass effect on surface hardness and hardness penetration with little concern as to grain size. Freedom from shape change also becomes predictable and an important factor.

With air hardening steels, again the question is what is the surface hardness obtainable for a given mass. In these steels, the hardness penetration is generally ample and, therefore, not an important consideration. A moderately high center hardness is generally desirable in both oil and air hardening steels due to application in order to give uniform, distortion free hardening and so that a tool may be reground many times without rehardening.

This paper deals directly with the properties of the more com-

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mon alloy and carbon tool steels and shows the user how these phenomena affect the tool steel he is using. Because of the large number of compositions offered and the many alloys employed plus the significance of this data, there seems to be a definite need for the record of hardenability figures made in accordance with accepted standard tests covering the more important tool steel classifications.

It is with these thoughts in mind that this paper has been prepared in such a manner as to aid the tool steel user in better selection of suitable steels for specific purposes.

TYPES OF TOOL STEEL STUDIED

The principal types of tool steels covered by this paper are shown in Table I.

Table I
Type and Typical Analysis of Tool Steels Studied

Type	Typical Analysis					
	C	Mn	Si	W	Cr	Va
Carbon	1.00	0.25	0.25
Carbon (Increased Manganese)	1.00	0.35	0.25
Carbon-Vanadium	1.00	0.25	0.25	0.20
Carbon-Chromium	1.00	0.25	0.25	0.20
Carbon-Chromium	1.00	0.25	0.25	0.45
Tap Steel	1.20	0.30	0.30	1.00	0.25
Fast Finishing	1.30	0.30	0.30	3.25	0.50
Shock Resisting	0.50	0.25	0.25	2.25	1.25	0.25
Nondeforming (Mn-Cr-W)	0.90	1.10	0.30	0.50	0.50	0.10
Nondeforming (Cr-W)	0.90	0.60	0.30	0.50	1.60
High Carbon, High Chromium	2.10	0.30	1.00	1.00	10.50
High Speed	0.70	0.30	0.30	18.00	4.00	1.25

From this table, it is evident that a wide range of compositions are available and that many alloying elements are employed in developing a variety of properties.

SPECIMEN PREPARATION

To demonstrate the hardening characteristics of these steels, all pieces were machined from bars $3\frac{1}{8}$ inches in diameter as shown on the sketch of Fig. 1. These samples were then cut to lengths making four samples as follows:

- 1 inch diameter by 3 inches long
- $1\frac{1}{2}$ inches diameter by 4 inches long
- 2 inches diameter by 5 inches long
- 3 inches diameter by 6 inches long

All samples previous to machining were given the standard mill anneal for the grade.

On heating for hardening, all samples were preheated 1 hour at 1200 degrees Fahr. (650 degrees Cent.) in an electric resistance type furnace having a hearth area of approximately 12 by 18 inches after which they were transferred to a furnace maintained at the

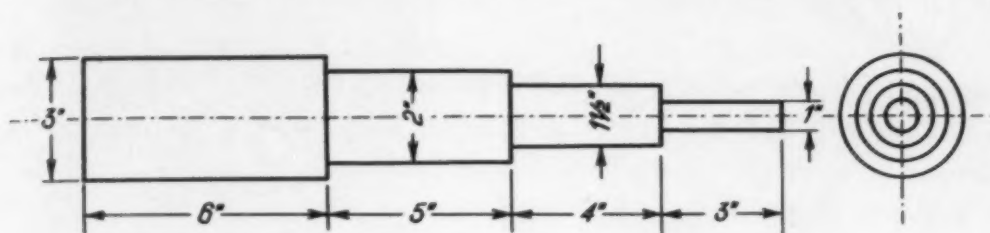


Fig. 1—Hardenability Test Specimen.

quenching temperature. All samples with the exception of those of high speed steel were held at temperature as follows:

1	inch	—	30 minutes
1½	inches	—	45 minutes
2	inches	—	60 minutes
3	inches	—	90 minutes

The water hardening steels were quenched in a still brine solution containing one pound of salt per gallon of water. The brine temperature did not exceed 80 degrees Fahr. The oil hardening samples were quenched in circulating, luke warm oil. Throughout this demonstration of hardenability, shop furnaces and equipment were used. No special quenching devices were employed. The water-quenched specimens were held in the quench by suitable tongs and kept in motion until below 200 degrees Fahr. The oil-quenched samples were set on a submerged screen in the agitated oil. The work was done in this manner with the thought that it would be representative of results that could be reasonably expected with average hardening room equipment.

After hardening, all the samples were notched and fractured. Half was photographed to show the macro grain size and hardness penetration. The other fractured face was ground smooth and Rockwell readings taken across the diameter and the results plotted. To further demonstrate the hardness penetration obtained from the water-quenched steels, several specimens were etched in a hydrochloric acid solution to bring out the hardness penetration.

In the water hardening steels, due to their dependence on high

quenching speeds, and importance of hardness penetration, it is pointed out that small composition variations visibly alter the depth of hardness. This difference is of particular value on many specific problems.

CHARACTERISTICS OF 1.00 PER CENT CARBON TOOL STEEL

Fig. 2 shows the hardenability curves of a 1.00 per cent straight carbon tool steel quenched in brine from 1450 and 1600 degrees Fahr. (790 and 870 degrees Cent.) for the sizes indicated. As is expected, the surface hardness in all instances is approximately 66 Rockwell. The depth of hardness obtained is shown photographically in Fig. 3.

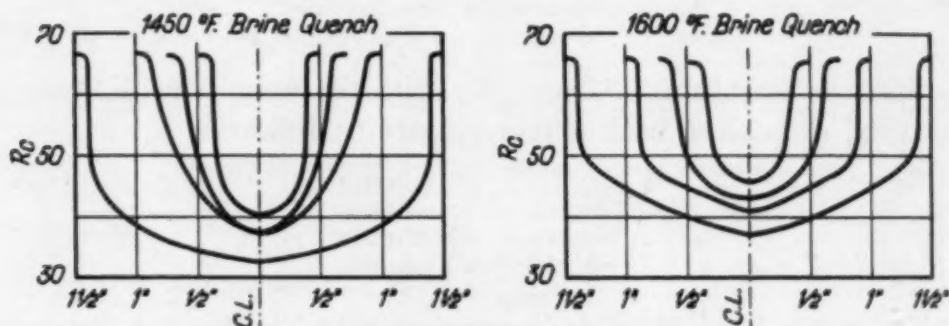


Fig. 2—Effect of Mass on Hardness Penetration. Carbon Tool Steel—Carbon 1.00 Per Cent.

Their fracture on quenching from 1450 degrees Fahr. (790 degrees Cent.) shows in hardened zone approximately No. 8 Shepherd grain size and from 1600 degrees Fahr. (870 degrees Cent.) approximately No. 6. The increased quenching temperature raises the core hardness approximately 5 points Rockwell and there is a slight increase in hardness penetration. The relatively soft core in water hardening steels is most valuable in many operations because of the shock dampening effect in combination with the high surface hardness. This gives in these steels both shock resistance and high abrasion resistance. It should be noted that in all instances there is a slight decrease in hardness penetration with increasing diameter of the specimens. For most applications of this type of steel, the user is more concerned with the results obtained with hardening at approximately 1450 degrees Fahr. (790 degrees Cent.). From the photographs, it can be seen that at this temperature good hardness penetration is obtainable; also, there is less difficulty due to decarburization of the work and a minimum of shape change for the type.

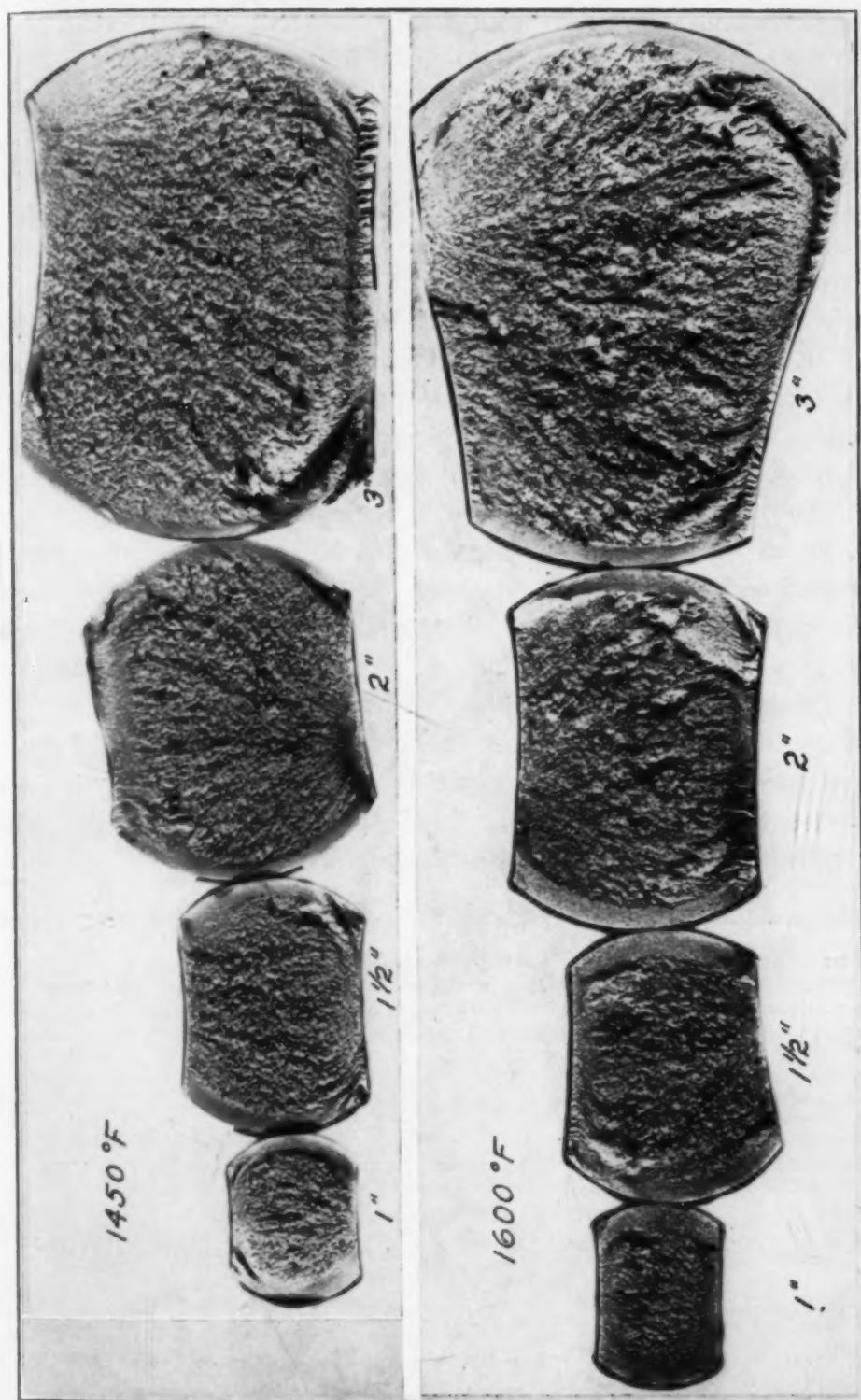


Fig. 3—Fracture Grain Size of Brine-Quenched 1.00 Per Cent Carbon Tool Steel.

CHARACTERISTICS OF 1.00 PER CENT CARBON TOOL STEEL WITH INCREASED MANGANESE

Small additions of the carbide forming elements—manganese, vanadium, chromium and tungsten—cause valuable variations of hardness penetration with only slight variation of macro grain size on carbon tool steels. As can be observed from the following hardness penetration charts and the fracture tests, the core hardness is not greatly disturbed; however, the hardened zone shows definite differences as to depth. To the uninitiated, these variations may appear small, but they are very useful on many types of operations. To illustrate, for large header die operations, requiring this class of steel, some increased depth of hardness over the straight carbon steel is desirable and is obtained by slightly increasing the manganese content to approximately 0.35. This will produce hardness to a greater depth and with increased uniformity particularly around the contour of the impression. This slightly increased depth of hardened zone offers greater resistance to sinking and cracking in the corners of these die impressions.

Figs. 4, 5, and 6 illustrate the effectiveness of the slightly increased manganese content. Table II also shows the relative hardness penetration in inches of the straight carbon steel and the carbon steel with slightly increased manganese.

Table II
Effect of Slight Increase in Manganese on 1.00 Per Cent Carbon Tool Steel

	1450° F. Brine Quench		1600° F. Brine Quench	
	C-1.00, Mn-0.25	C-1.00, Mn-0.35	C-1.00, Mn-0.25	C-1.00, Mn-0.35
1 inch diameter	0.1094	0.1250	0.1561	0.2500
1½ inches diameter	0.0937	0.1094	0.1250	0.2182
2 inches diameter	0.0781	0.1094	0.1250	0.1719
3 inches diameter	0.0625	0.0781	0.0937	0.1561

CHARACTERISTICS OF 1.00 PER CENT CARBON TOOL STEEL WITH VANADIUM

Vanadium additions tend to reduce the depth of the hardened zone (Figs. 7, 8, and 9) but the macro grain size is exceedingly fine and experience has shown that this type of steel has a very hard, long wearing and elastic hardened zone which with the soft core makes it desirable for punches and dies subject to severe abrasion

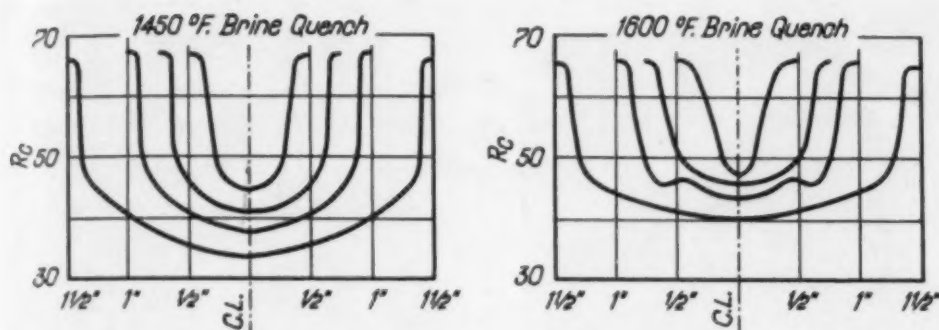


Fig. 4—Effect of Mass on Hardness Penetration. Carbon Tool Steel—Manganese 0.35 Per Cent.

and impact. Table III illustrates the difference in hardness penetration between a 1.00 per cent carbon steel without vanadium and one containing 0.20 vanadium.

Table III
Effect of Vanadium on 1.00 Per Cent Carbon Tool Steel

	1450° F. (790° C.)		1500° F. (815° C.)	
	C-1.00	C-1.00, V-0.20	C-1.00	C-1.00, V-0.20
1 inch diameter	0.1094	0.0937	0.1250	0.1094
1 1/2 inches diameter	0.0937	0.0781	0.1250	0.1094
2 inches diameter	0.0781	0.0781	0.0937	0.0781
3 inches diameter	0.0625	0.0781	0.0937	0.0625

CHARACTERISTICS OF 1.00 PER CENT CARBON TOOL STEEL WITH CHROMIUM

Figs. 10, 11, 12, 13, and 14 and Table IV show the effect of increased chromium additions on hardness penetration of 1.00 per cent carbon steel. The increased depth of the hardened zone which occurs in the 0.20 per cent chromium steel is desirable for noble metal dies and silverware dies when resistance to sinking is a problem. Large size blanking, forming and drawing dies of the 0.45 per cent chromium composition harden with increased uniformity and with a high surface hardness which also penetrates to a considerable depth, offsetting the effect the mass would have on straight carbon tool steel. Also, if the total number of pieces to be produced is large, in many types of tools increased regrindings are possible with both of these chromium steels before it is necessary to reharden. Further, in all instances, it appears that the addition of chromium somewhat increases the wearing properties of carbon steels.

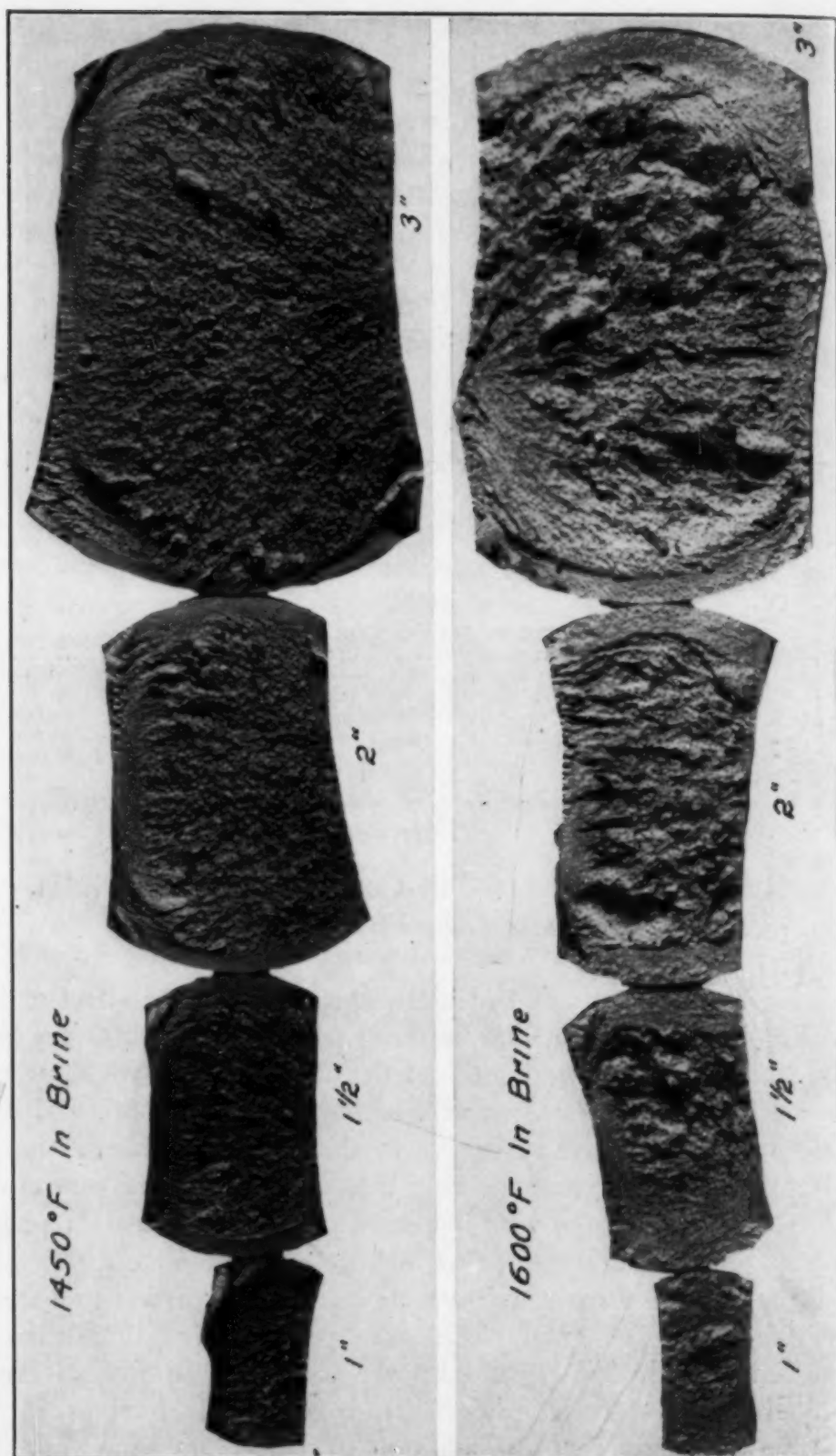


Fig. 5—Fracture Grain Size of Quenched Carbon Tool Steel With 0.35 Per Cent Manganese.

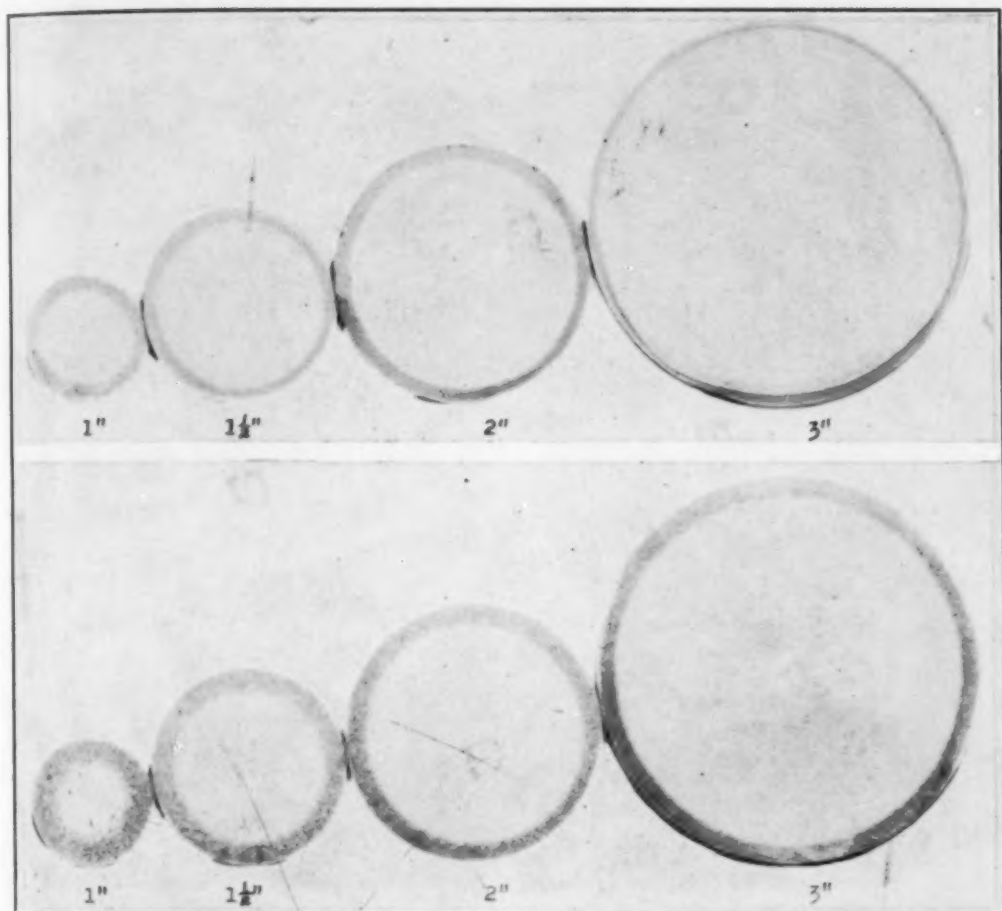


Fig. 6—Effect of Manganese on Hardness Penetration Quenched From 1600 Degrees Fahr. in Brine. Upper—1.00 Per Cent Carbon Tool Steel. Lower—1.00 Per Cent Carbon Tool Steel With 0.35 Per Cent Manganese.

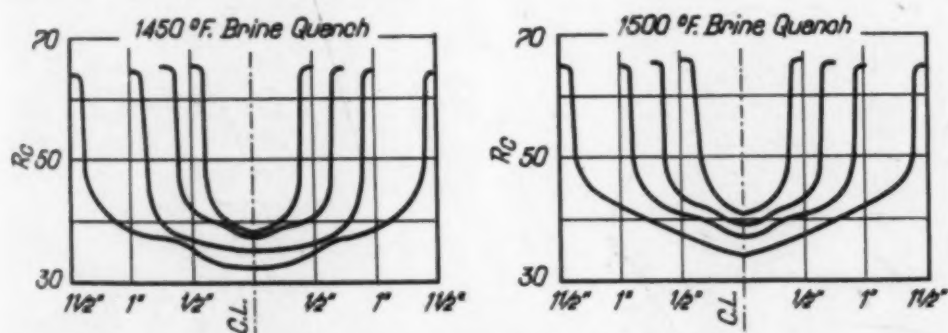


Fig. 7—Effect of Mass on Hardness Penetration. Carbon Tool Steel—V 0.25 Per Cent.

Table IV
Effect of Chromium on 1.00 Per Cent Carbon Tool Steel

	1450° F. (790° C.)			1500° F. (815° C.)		
	C-1.00	C-1.00, Cr-0.20	C-1.00, Cr-0.45	C-1.00	C-1.00, Cr-0.20	C-1.00, Cr-0.45
1 inch diameter	0.1094	0.1250	0.2500	0.1250	0.1250	0.2500
1 1/2 inches diameter	0.0937	0.1250	0.1719	0.1250	0.1250	0.1875
2 inches diameter	0.0781	0.1250	0.1405	0.0937	0.1250	0.1719
3 inches diameter	0.0625	0.1095	0.1405	0.0937	0.1250	0.1405

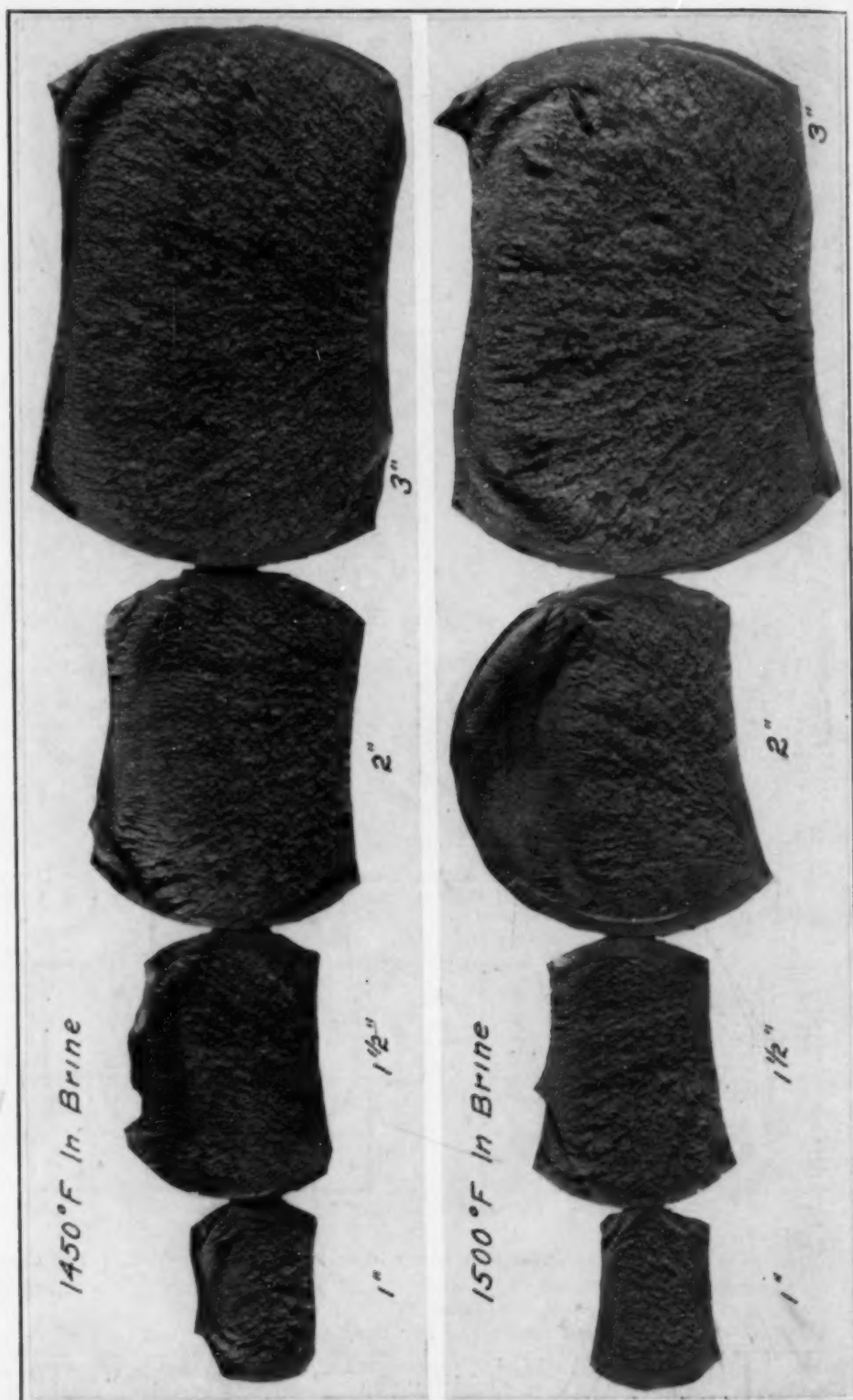


Fig. 8—Fracture Grain Size of Quenched Carbon-Vanadium Tool Steel.

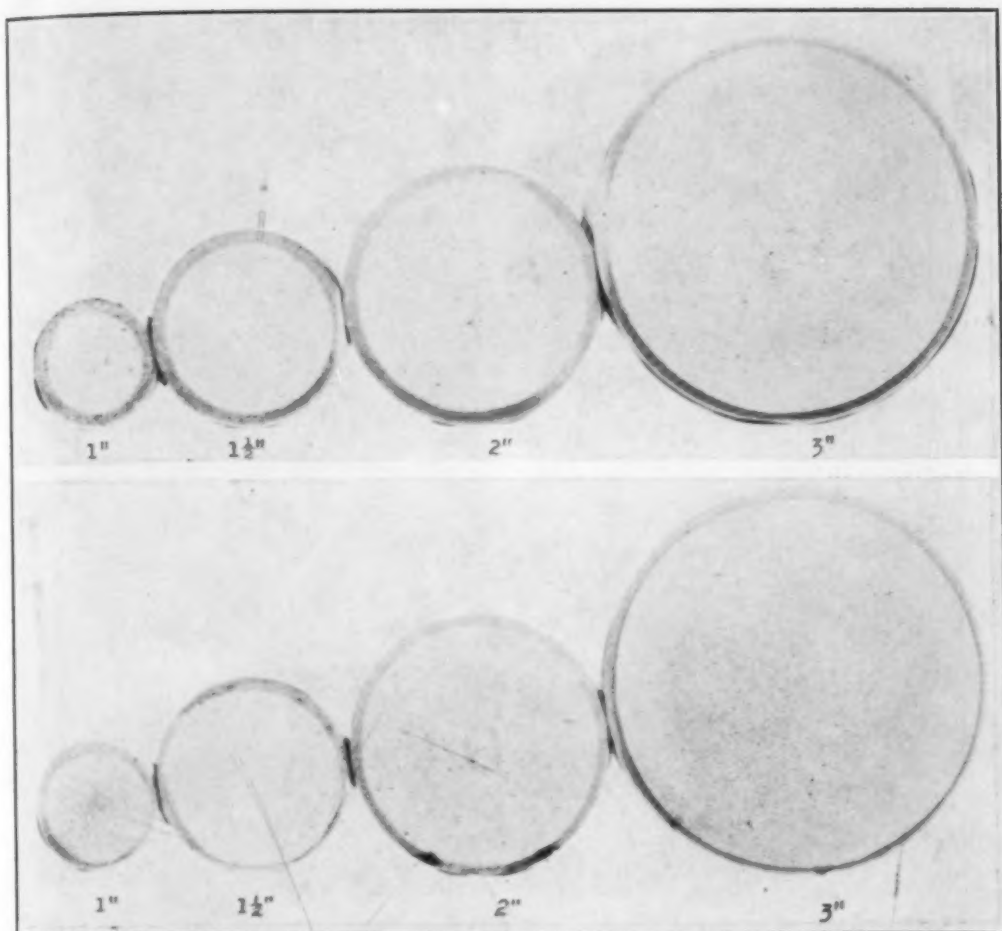


Fig. 9—Effect of Vanadium on Hardness Penetration Brine-Quenched From 1450° F. Upper—1.00% C. Lower—1.00% C, 0.20% V.

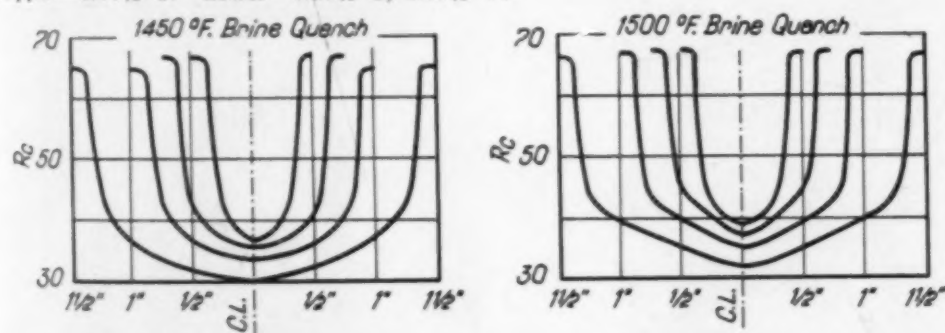


Fig. 10—Effect of Mass on Hardness Penetration. Carbon Tool Steel—Cr 0.20%.

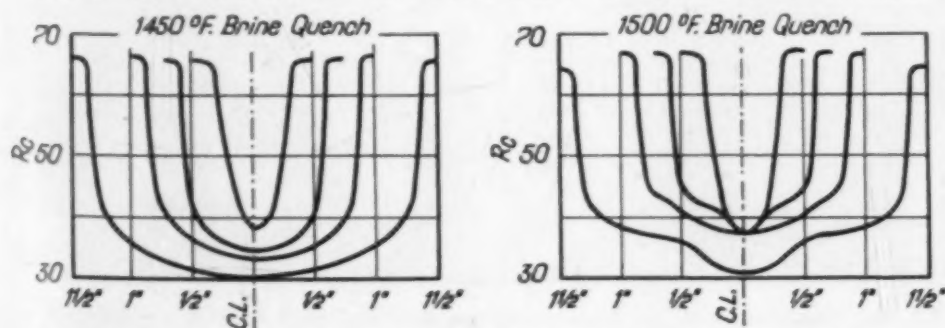


Fig. 11—Effect of Mass on Hardness Penetration. Carbon Tool Steel—Cr 0.45%.

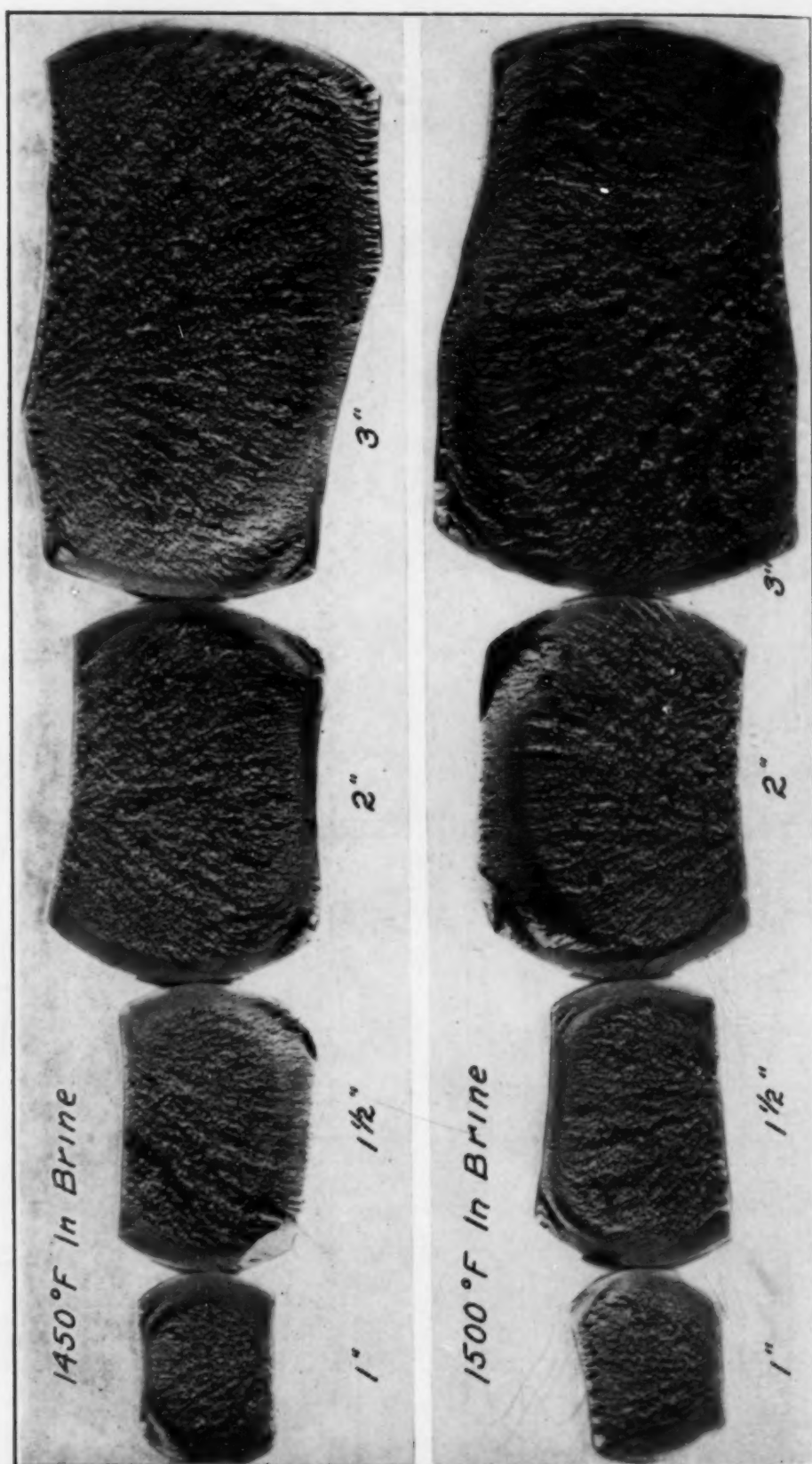


Fig. 12—Fracture Grain Size of Carbon-Chromium Tool Steel (0.20 Per Cent Cr).

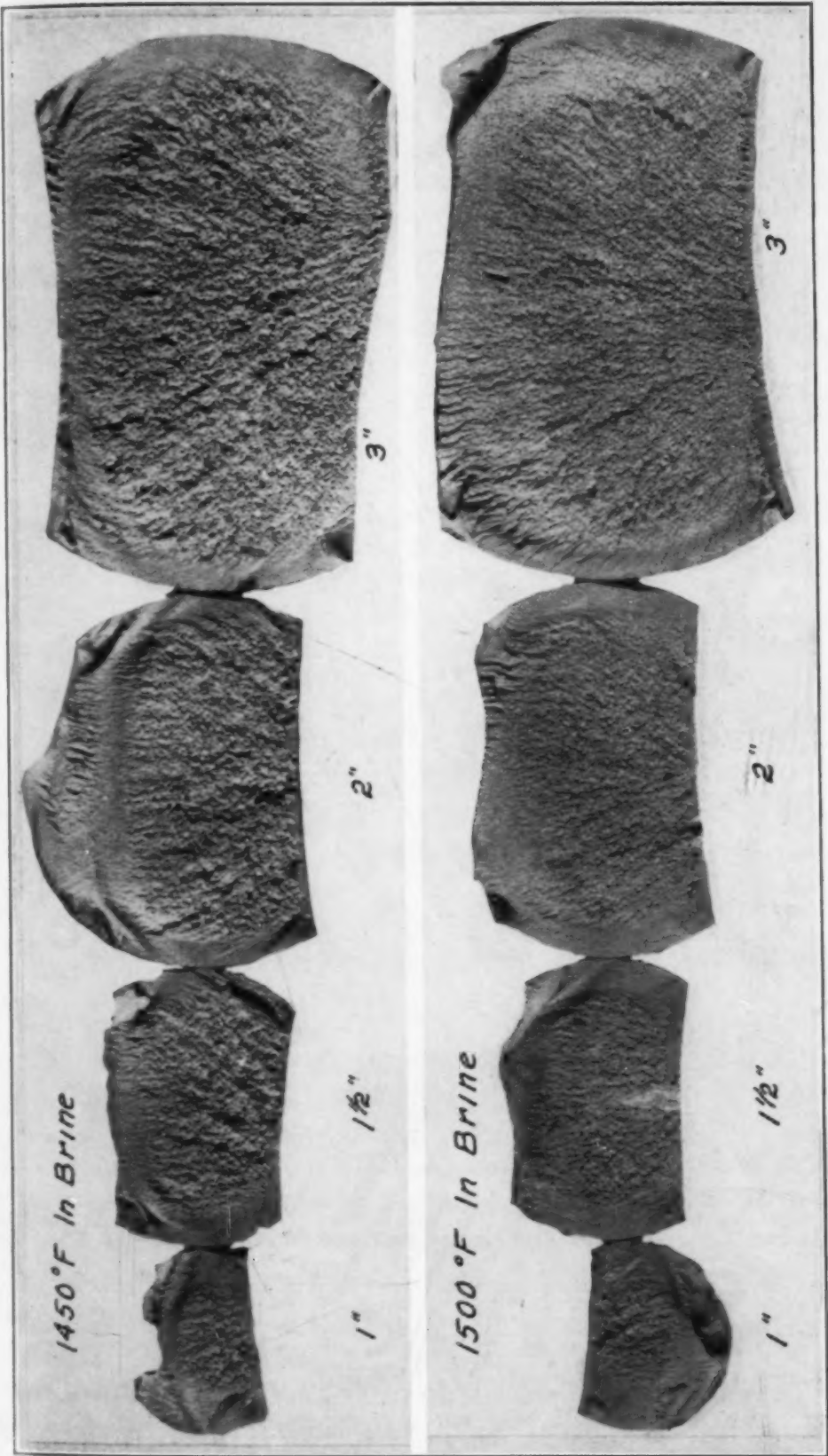


Fig. 13—Fracture Grain Size of Carbon-Chromium Tool Steel (0.45 Per Cent Cr).

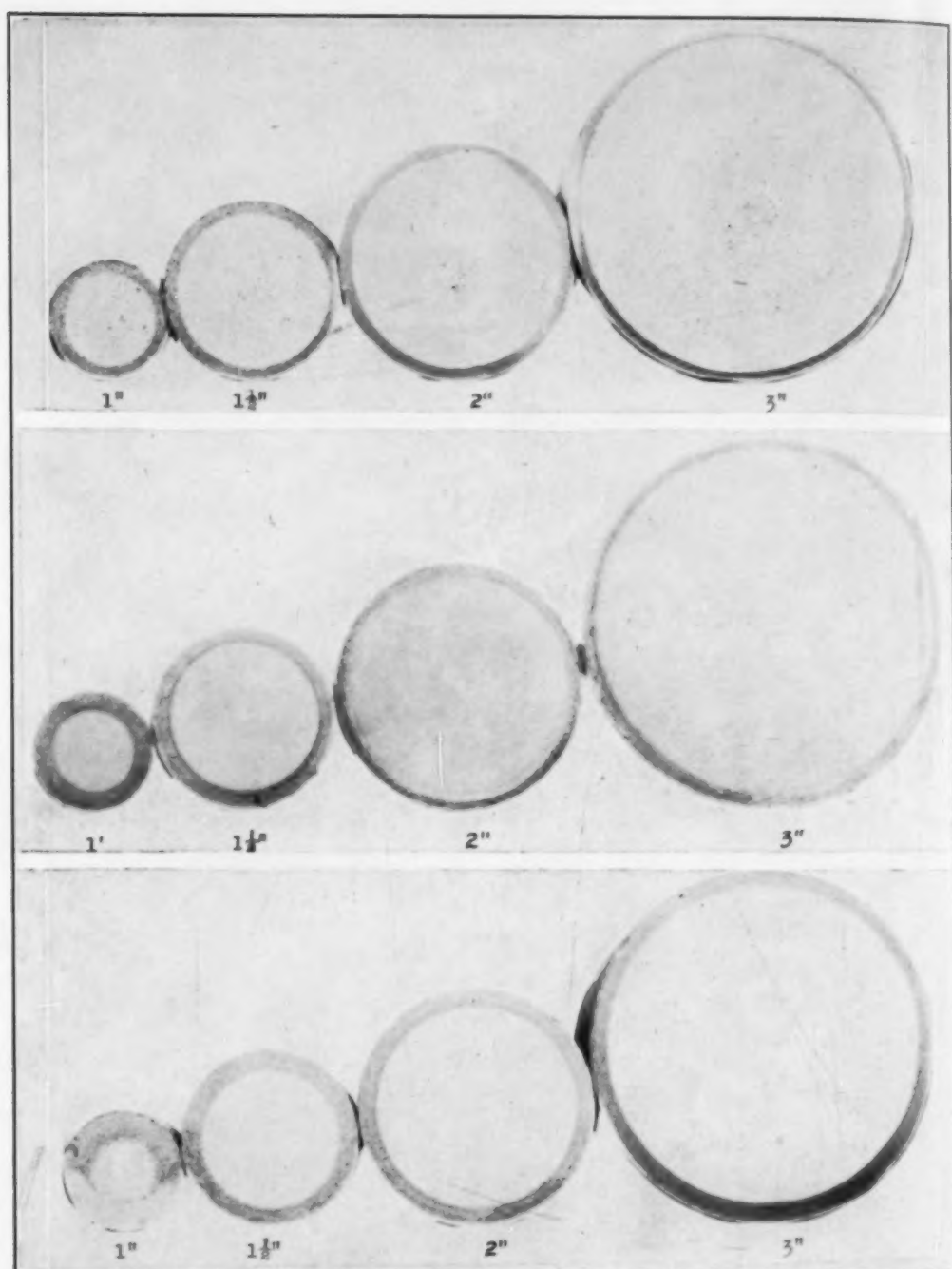


Fig. 14—Effect of Chromium on Hardness Penetration Quenched From 1450 Degrees Fahr. in Brine. Upper—1.00 Per Cent Carbon Tool Steel With 0.20 Per Cent Chromium. Lower—1.00 Per Cent Carbon Tool Steel With 0.45 Per Cent Chromium.

CHARACTERISTICS OF TAP AND FAST FINISHING STEELS

Tungsten alone does not appreciably affect the depth of hardness of water hardening steels but it does increase wearing ability and

when added in quantities of 1 per cent it produces high hardness and keen cutting ability.

Figs. 15 and 16 show the hardness penetration and fracture grain size of a tungsten tap steel. Steels of this type are usually water hardening. This tap steel also is relatively shallow hardening but because of the keenness of cutting developed by the tungsten, it is very useful for many tapping operations.

Figs. 17 and 18 show the hardness penetration and fracture grain size of the fast finishing type tool steel. Steels of this type

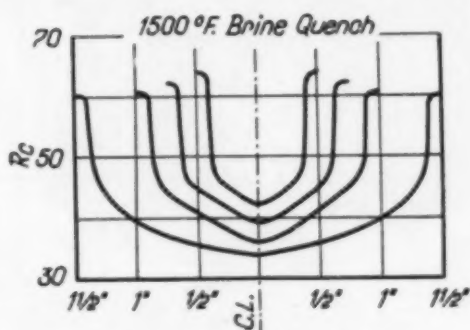


Fig. 15—Effect of Mass on Hardness Penetration of Tap Steel.

are water hardening. The fast finishing steel has a very fine fracture grain size. This steel with its extremely high surface hardness and unusually keen cutting ability with some increase in depth hardness is particularly useful for cutting nonferrous products and machining chilled iron rolls.

CHARACTERISTICS OF SHOCK RESISTING AND NONDEFORMING STEELS

In the higher hardenability steels suitable for oil quenching, we find the fracture appearance changed in that there is not the definition of hardened zone and soft core observed in the water hardening steels. The fracture appearance is very fine throughout the cross section and the photograph in Fig. 19 is the characteristic appearance of most of the oil hardening steels and high speed steel.

The hardness variation from surface to center is greatly reduced. The value of these steels lies in the fact that intricate shapes made from these steels can be quenched with a minimum of danger of cracking and shape change. Because of the low critical cooling speed necessary, they harden uniformly as to hardness and volumetric displacement. These steels have a wide application for tool and die



Fig. 16—Fracture Grain Size of Quenched Tungsten Tap Steel.

work by reason of their freedom from shape change and permissible regrindings without rehardening.

In these oil hardening steels, the elements—manganese and chromium—are added to impart high hardness. Tungsten is added to increase the wear resistance of the steel and does not greatly influence its hardenability. Fig. 20 shows the characteristic hardness penetration of the shock resisting, oil hardening tool steels. This composition with its low surface hardness but with deep hardening qualities is very useful for shock tools not subject to severe abrasion. It will carry heavy loads and resist heat checking when used for hot work applications.

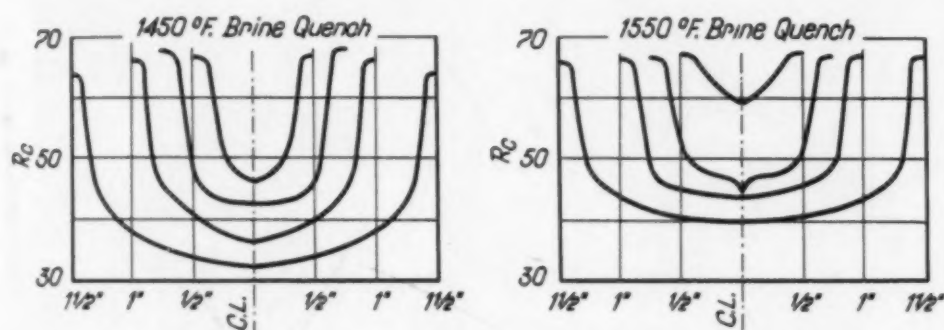


Fig. 17—Effect of Mass on Hardness Penetration of Fast Finishing Steel.

Figs. 21 and 22 show the hardness penetration of the Mn-Cr-W and the Cr-W type nondeforming, oil hardening tool steels. These steels harden very uniformly with little tendency to distort because of their uniformity of hardness and displacement throughout the mass. Their accuracy in hardening makes them useful for intricate tools where cutting edges cannot be reground after hardening. They will outlast the water hardening steels because of permissible surface regrinding. These steels are suitable for all types of cold die work.

CHARACTERISTICS OF HIGH CARBON, HIGH CHROMIUM STEELS

This type of steel is the outstanding die steel for many applications such as blanking, forming, trimming, etc. The hardenability curves (Fig. 23) show the highest hardness obtainable of any of the die steels throughout the section on oil quenching. This steel will also air harden in substantial sections. It is nondeforming and has the highest hardness and greatest wear resistance of any of the die steels. Considerable hardness is retained at temperatures up to 1000

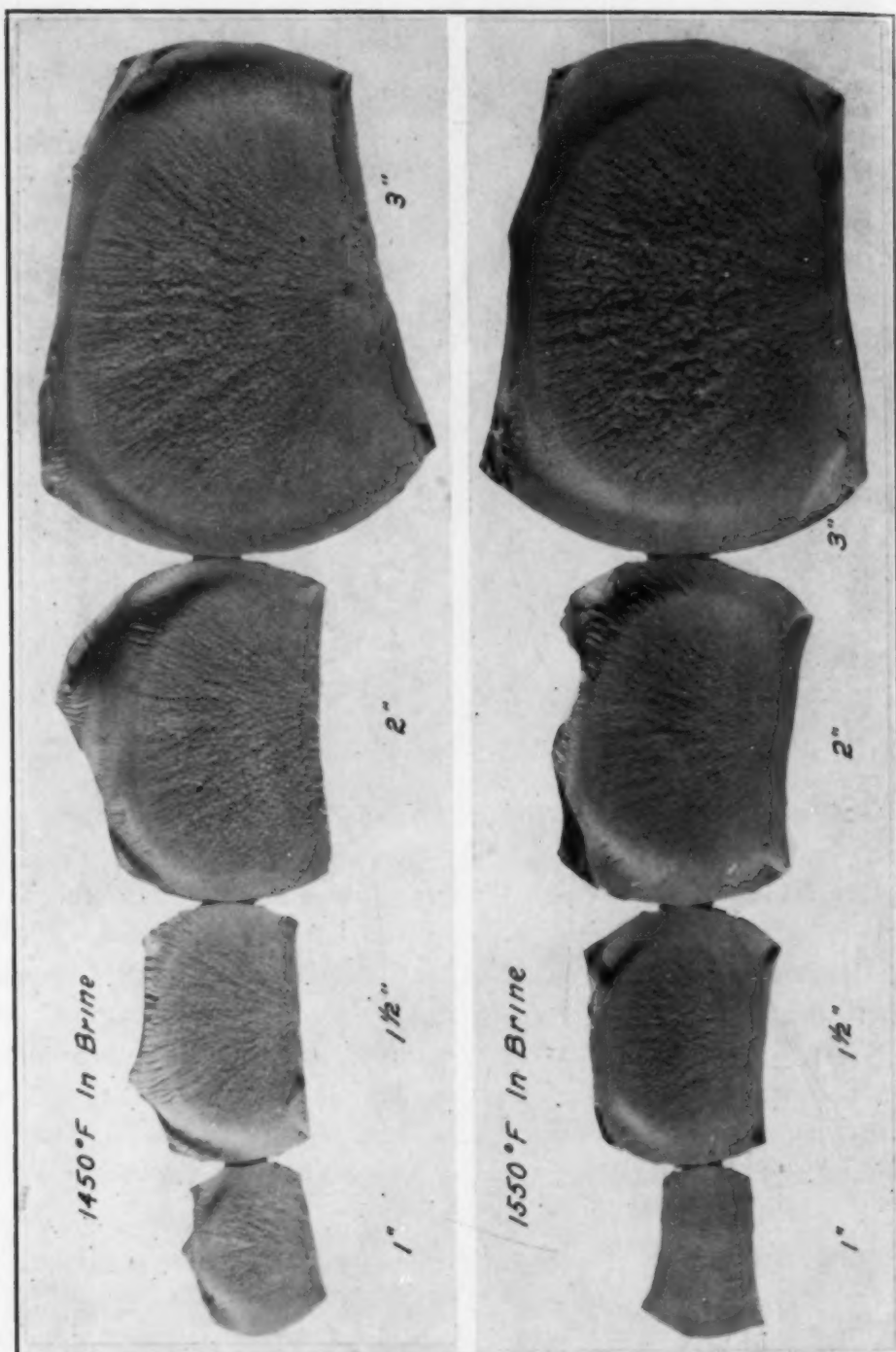


Fig. 18—Fracture Grain Size of Quenched Fast Finishing Steel.

degrees Fahr. While these steels have relatively low impact strength, they have exceedingly high crushing strength and, due to their deep hardening, they show almost no change in wear resistance as the surfaces are reground.



Fig. 19—Characteristic Fracture Appearance of Oil Hardening Die Steel.

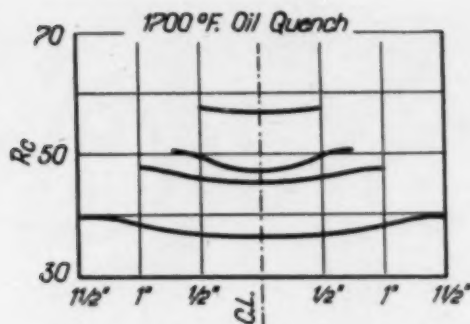


Fig. 20—Effect of Mass on Hardness Penetration of Shock Resisting Tool Steel.

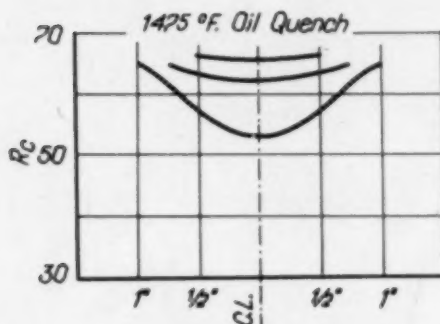


Fig. 21—Effect of Mass on Hardness Penetration Nondeforming (Mn-Cr-W) Tool Steel.

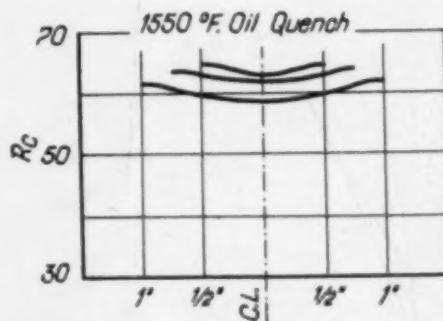
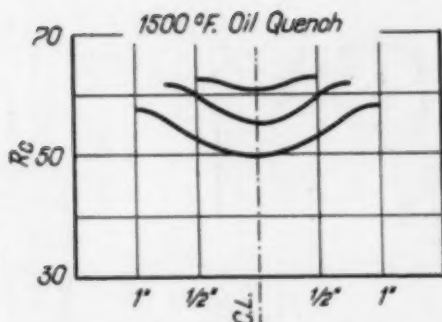


Fig. 22—Effect of Mass on Hardness Penetration Nondeforming (Cr-W) Tool Steel.

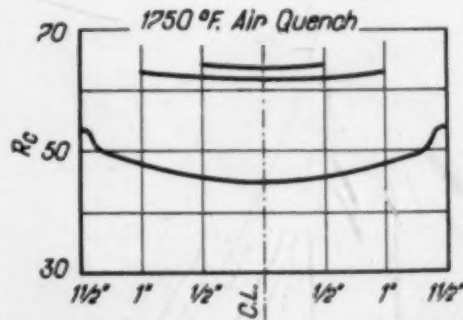
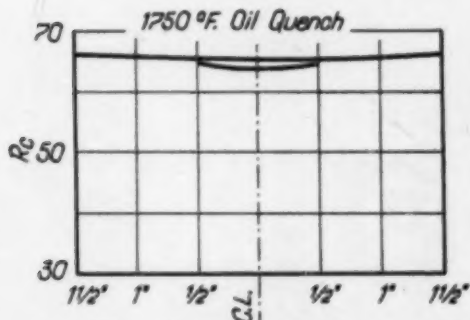


Fig. 23—Effect of Mass on Hardness Penetration of High Carbon-High Chromium Steel.

CHARACTERISTICS OF HIGH SPEED STEEL

To demonstrate the hardenability of high speed steel, specimens were preheated in semi-muffle, gas fired furnaces for 45 minutes at 1100 degrees Fahr. and 45 minutes at 1600 degrees Fahr. They were then transferred to a high temperature, semi-muffle gas furnace maintained at 2350 degrees Fahr. (1290 degrees Cent.). Their holding time in the high heat furnace was as follows:

- 1 inch diameter — 4½ minutes
- 1½ inches diameter — 6½ minutes
- 2 inches diameter — 8 minutes
- 3 inches diameter — 10½ minutes

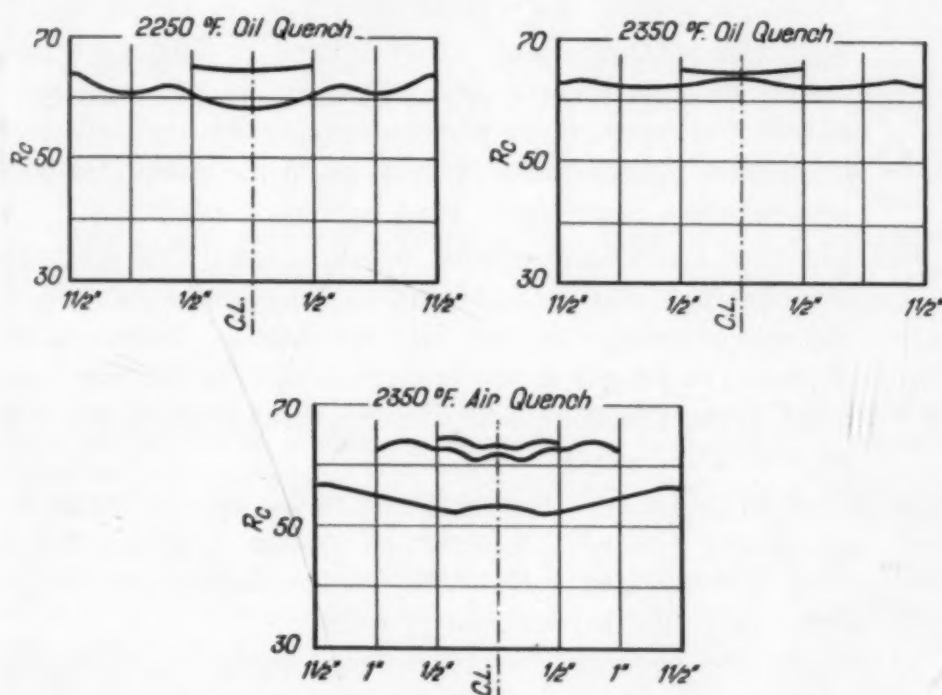


Fig. 24—Effect of Mass on Hardness Penetration of High Speed Steel.

High speed steel is the outstanding cutting steel of all the tool steel types. Its characteristic hardenability curves are shown in Fig. 24. Like the high carbon, high chromium steels, it develops high hardness as quenched and is both oil and air hardening. It has the greatest cutting capacity of any of the tool steels, because of its red hardness properties due to the tungsten which results in the retention of high hardness up to 1100 degrees Fahr. As a type, it has the highest, all-around physical properties of any of the tool steels. It has the highest tensile and compressive strength although the impact strength is only moderate. Its abrasion resistance is high and is

retained even in large sections after many regrindings. While the high hardening heat introduces a new problem in the hardening, it is relatively free from distortion due to its uniformity of hardening.

CONCLUSIONS

1. As we move "up" in the established types of tool steel from the carbon steels to the higher performance and safer steels, the difference in the steels is explained by the difference in surface hardening effect plus hardenability (or depth hardening) until high speed steel is reached where red hardness is an added quality.
2. From the hardenability values shown, it is apparent that:
 - (a) *Standard tensile and compression test results*, which in small size specimens are closely related to hardness values, cannot indicate the physical properties in tool size specimens. However, these physical properties are apparently predictable in tool size specimens from the composite hardness readings taken from hardenability curves.
 - (b) *Standard impact test results*, which in small size specimens are closely related to hardness values, cannot indicate the physical properties in tool size specimens. However, this physical property is apparently predictable in tool size specimens from the composite hardness readings taken from hardenability curves.
 - (c) *Tool life* (for initial grinds, regrinds, and ultimate performance) which is a function of wear resistance which, in turn, is related to indentation hardness tests, can be predicted from the hardenability curves.
 - (d) *Distortion*, commonly evidenced by warpage and shrinking, resulting from quenching tool steel and caused by the smaller displacement of material of reduced hardness can be predicted from hardenability curves.
3. From these indications, we believe that the hardenability test when applied to tool steel permits more intelligent selection than all the other test methods ordinarily considered.

DISCUSSION

Written Discussion: By H. T. Morton, chief metallurgist, and E. E. Wagner, assistant metallurgist, Hoover Ball Bearing Co., Ann Arbor, Mich.

Mr. Scheid is to be congratulated on his hardenability tests covering tool steels as most work has been done on production steels. The most interesting

results are shown in Table II where a change from 0.25 per cent to 0.35 per cent manganese increases the hardness penetration on an average of 25 per cent at 1450 degrees Fahr. and 59 per cent at 1600 degrees Fahr. If this result remains consistent after checking numerous heats, it might become a determining factor in the selection of deep hardening tool steels for heavy working tools or dies, or shallow hardening steels for light work and toughness.

It would be desirable if Mr. Scheid had given more details as to his technique in heating and quenching his samples, as well as measuring them.

Written Discussion: By C. L. Harvey, chief metallurgist, The Lamson and Sessions Co., Cleveland.

I feel that the author should be complimented for this interesting paper on the hardenability of tool steel. Practically every type of steel commonly used has been covered by this research. However the picture would be more complete if data were given for the so-called air hardening high carbon high chromium type having approximately 1.5 per cent carbon, 12.0 per cent chromium and 1.0 per cent molybdenum. It would be interesting to compare the curve for this steel with the oil hardening type shown in Fig. 23.

I should also like to ask the author what spread in values shown in Tables II, III, IV might be expected when different heats of steel from the same mill were tested?

While I quite agree that hardenability characteristics under controlled heating and quenching cycles are of great value in the selection of steel for a given purpose, I cannot agree that physical properties such as tensile, compression and impact can be predicted from these curves except in a very broad sense. If the author is able to make such predictions accurately, I should certainly like to know his method. It has been my experience that hardenability curves can be used as a guide in the selection of steels but that they must be tied in with actual service tests for best results.

Written Discussion: By Victor Stefanides, metallurgist, Illinois Tool Works and Shakeproof Lock Washer Co., Chicago.

We are indebted to the author for selecting a subject for his research which is one of the most difficult problems metallurgists are confronted with when dealing with the design of dies for various applications.

In selecting a steel for a specific industrial problem, be it a cold header, trimming, forming or other type of die, the problem would, indeed, be simplified if the metallurgist could consult chemical analysis tables of some typical tool steel, with their "hardenability" characteristics, to make his final choice. Unfortunately, steels are not so simple, because each one has a temperament of its own, requiring the utmost care to realize the most out of it.

The author is doing an invaluable service to the industries by pointing out the effect of some of the alloying elements on the "hardenability" characteristics, yet it is to be regretted that other factors, which are very potent in their influence on the "hardenability," are not presented in this paper. To make this point clear, let us take, for the sake of illustration, the carbon (1.0 per cent carbon, 0.25 per cent manganese, 0.25 per cent silicon) steel in Table I, which according to the data in Table II, when quenched from 1450 degrees Fahr. gave 0.1094-inch depth of case, and when quenched from 1600 degrees Fahr. gave 0.1561-inch depth of case, for 1-inch diameter. The writer has quenched steels of identical

analysis from 1550 to 1600 degrees Fahr. and found no hardness variation from surface to center. It is conceded that such a condition is rather an exception than a rule. However, it serves to illustrate that one set of conditions will give one result, whereas slight variation in a given set of conditions may give an entirely different result.

The tool steel metallurgists are familiar with the deep hardening characteristics of incompletely degasified 1.00 per cent carbon steels. The deep hardening characteristic is, of course, due to the presence of oxygen. The coarse-grained steel, due to its incomplete degasification, is rendered coarse and, as such, lacking the Al_2O_3 nuclei acting as centers for crystallization, will also harden deeper than fine-grained steel. Whether it is the presence of oxygen in the coarse-grained steel, or lesser number of the Al_2O_3 particles acting as nuclei for the decomposition of austenite at the grain boundaries, is not clear, yet the fact remains that the presence of oxygen or coarse grain facilitates the depth of case.

It is also a well known fact that the 1.0 per cent carbon steel with vanadium addition does not harden to the same depth as the same steel without the vanadium. Vanadium, because of its potent scavenging action, reduces the oxygen to a minimum. The vanadium carbides, which form in preference to iron carbides, act as a deterrent to grain growth, because they do not form solid solution with the gamma iron at the ordinary heat treating temperatures. Furthermore, on quenching, the vanadium carbides may act as nuclei for austenite decomposition. However, the writer believes that 1.0 per cent carbon steel free of oxygen will harden to the same depth as the same steel with the vanadium addition, providing other variables remain the same.

Because of the harder nature of the vanadium carbides over that of iron carbides, when present in 1.0 per cent carbon steel should improve its wear resisting properties.

The writer does not attach much importance to the presence of a small amount of manganese or chromium. The depth of hardness can, of course, be increased by their presence; however, their value is very doubtful as far as the wear resisting properties are concerned. Chromium, especially when present in small quantities, forms Cr_7C_3 carbides, which does not add materially to the quality of the steel.

With reference to the nondeforming steels, it is surprising how the so-called nondeforming steels will deform on heat treating, especially the steels which on macro etching will exhibit the well known open structure. However, the deforming characteristic can be reduced to a minimum by drop forging practice.

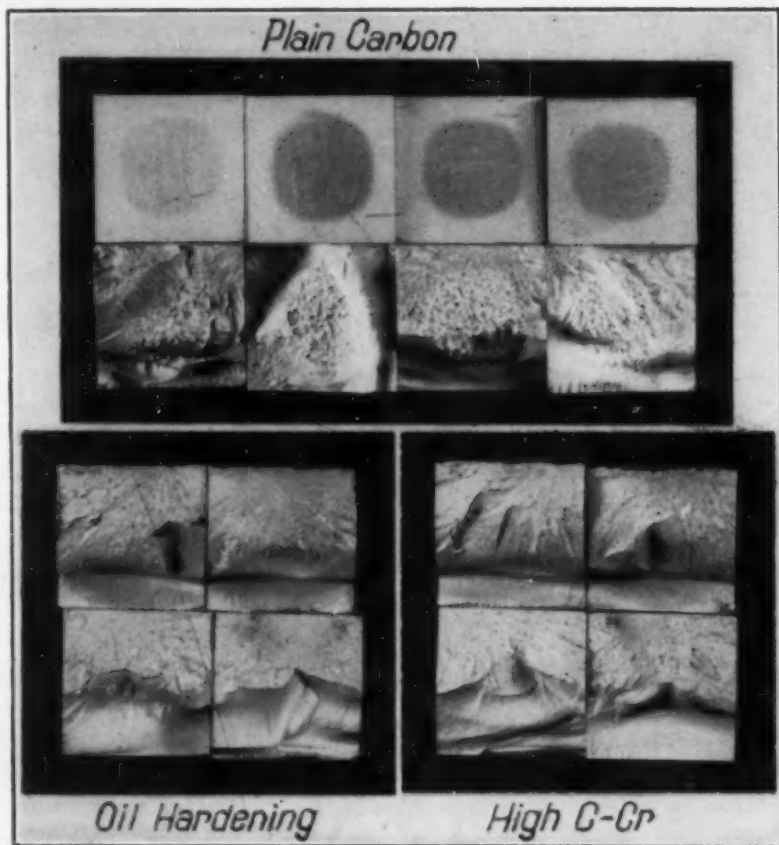
Written Discussion: By John E. Erb, laboratory metallurgist, General Electric Co., Schenectady, N. Y.

Next to its chemical composition, the most important thing to know about a tool steel is its hardenability. Mr. Scheid has, therefore, rendered a splendid service in providing us with data on the hardening characteristics of tool steels.

Along with these useful results, however, are certain broad statements and conclusions, which, to the present writer, seem to have little connection with the data.

For example, on page 448 the statement is made, "The relatively soft

core in water hardening steels is most valuable in many operations because of the shock dampening effect in combination with the high surface hardness. This gives in these steels both shock resistance and abrasion resistance." Admitting that the presence of a core does improve the shock resistance of carbon steel, there also seems to be a widespread impression, often appearing in print, that carbon steel is shock resistant as compared with other types of tool steel. On our part we find carbon steel rather poor in this respect, even in sections which harden with a "core." Evidence is offered in the accompanying table and photograph, showing the core in the carbon steel specimens.



McMullan¹ has also attacked this theory as applied to case hardening steels.

On page 466, under the second conclusion, it is stated that the physical properties of tool size specimens are apparently predictable from the composite hardness readings taken from hardenability curves. I would like to have Mr. Scheid explain in detail just how he proposes to do this.

To his final conclusion, I desire to register a violent protest. Because the hardenability test is a good one, it does not follow that all other tests are poor by comparison. Knowledge of strength, toughness, red-hardness, wear-resistance, and of many other things is badly needed today. Granted that certain "high-flown" aspects of the subject have received too much attention at the expense of basic ones such as hardenability; also granted that tests of doubtful value have

¹"The Properties of Case Hardened Steels," O. W. McMullan, TRANSACTIONS, American Society for Metals, Vol. 23, 1935, p. 319-381.

Table I
Tool Steel—Comparison of Strength and Toughness Tests

	Plain Carbon	Oil Hard	High C-Cr
1. Charpy Impact Unnotched Specimen 1/4 x 3/8 x 2 Inches Long	7.4	30	28
2. Izod Impact Unnotched Specimen 3/8 Inch Square x 3 Inches Long	68	86*	94*
3. Modulus Rupture Cantilever Bend Test 3/8 Inch Square x 6 Inches Long	298,000	346,000	148,000
4. Hardness Rockwell C	64	62	64
5. Type Composition:			
C	1.10	0.90	1.60
Mn	0.25	0.30	0.30
Si	0.20	0.30	0.50
Cr	1.60	12.00
W	0.45
Mo	1.00
V	0.25
Co	0.60
6. Heat Treatment			
Harden	1450 Degrees Fahr.	1550 Degrees Fahr.	1875 Degrees Fahr.
Quench	Brine	Oil	Air
Temper	350 Degrees Fahr.	350 Degrees Fahr.	300 Degrees Fahr.

*Probably low. Two specimens out of four didn't break.

received wide publicity, or that good tests have been misinterpreted, it ill serves the progress of an industry to disparage the well meant efforts in this direction.

Rather, in the writer's opinion, should the aims of this Society be devoted to fostering more and better tests, standardization of them, and promotion of their use by both makers and users of tool steel.

Oral Discussion

H. W. GRAHAM:² Personally I have been interested in this program. It has seemed to contain much of specific value, but as I listened I have been thinking of other types of hardness problems of a character such as we all face from time to time. For instance, we had a problem not long ago which related to abrasive wear of steel against geological structure. A geologist involved was very caustic in his comments about the steel industry using hardness tests such as Brinell and Rockwell and wanted us to use the Mohs' scale. I gave the matter some thought but the thought did not lead anywhere. We are occasionally brought to see that even the best of our hardness testing still leaves out of account something that is inherent in the character of steel. In lathe cutting and like operations we find some steels have run through a gamut of behavior or character that rivals the wide range of the physical quality of cheese. The optimum structure and characteristics that we usually struggle toward are of a dry crumbly nature. This is most helpful in machining, but no approach that we have been able to arrive at has ever led to a type of hardness measurement that would really portray the grain of the steel under survey.

²Director of metallurgy and research, Jones and Laughlin Steel Corp., Pittsburgh.

Author's Reply

With regard to Mr. Morton's comment on the use of further manganese additions for increased hardness penetration, I can readily see his point. However, if additional hardness penetration was obtained by increased manganese additions other undesirable factors would develop. With increased manganese, brittleness becomes an important factor and in most instances where increased depth of hardness is desirable it would be better to use chromium to obtain this increase in hardness penetration.

With regard to Mr. Stefanides' discussion of the 1.00 per cent carbon steels and the depth of hardness he has observed with steels of apparently identical analysis, the one factor I purposely left out of this discussion was the effect of aluminum. This subject has been well covered by others. In this outline of the typical tool steels the aluminum addition is a standard for the type. It is a constant as far as the particular type is concerned and the physical properties are altered by varying the percentage of the other alloying elements. It is true that the 1.00 per cent carbon steels, deoxidized with silicon only, will be deep hardening particularly when hardened from 1600 degrees Fahr. The illustration used in the paper was a medium grain straight carbon tool steel.

With reference to Mr. Harvey's question as to what spread in values shown in Tables II, III and IV might be expected: I would hesitate to set definite limits at this time; however, it is surprising how closely tests check on given compositions with a given mill practice.

In answer to Mr. Erb's violent protest, I wish to emphasize the fact that the object of this paper was to set forth a relatively simple test intended for practical use. The average tool steel user, in many instances, lacks the necessary laboratory equipment to determine the many standard physical tests and he is very much in need of a simple indicator as to the inherent properties of the various types of tool steel. We believe that this is best determined by the hardenability test. Mr. Erb's illustrations on $\frac{5}{8}$ -inch square samples are undoubtedly correct but are not at all representative of the physical properties that might be anticipated in larger sections.

DIMENSION CHANGES OF TOOL STEELS DURING QUENCHING AND TEMPERING

BY EINAR AMEEN

Abstract

The present paper describes an investigation which has been carried out to determine the dimension changes when quenching and tempering a number of various sized cylinders and rings made from unalloyed and alloyed tool steel.

When hardening, deformation takes place on account of heat stresses, as well as stresses which develop in connection with the transformation from austenite to martensite. Due to heat stresses a piece of steel tends to assume a shape approaching the spherical. Transformation stresses, on the other hand, cause the opposite kind of deformation.

The amount of deformation depends upon a number of factors such as velocity of cooling, elastic limit, heat conductivity and through hardening properties of the steel, amount of retained austenite and carbides present in the steel, shape of the piece, etc.

When tempering a piece of steel the dimensions change on account of structure transformations and also due to release of strains. Strain release seems to occur rapidly when tempering at about 200 degrees Cent. (390 degrees Fahr.) and thereafter slower.

Only in exceptional cases can we by tempering count upon complete elimination of deformation which has taken place during the hardening.

In order to determine the tendency of a steel to change dimensions during hardening, it seems advisable for small size tools to measure the change of the inner diameter when hardening a ring with outer diameter 50 millimeters, inner diameter 5 millimeters, and height 10 millimeters. For larger sized tools it is advisable to measure the change of the height of a cylinder of about 50-millimeter diameter and 50-millimeter height.

WHEN steel is heat treated there occurs, besides changes in the structure, changes in the volume and dimensions of the piece. The change in volume during heat treatment, which is caused

The author, Einar Ameen, is assistant to the technical director, Uddeholms A/B, Uddeholm, Sweden.

by structural changes, has been studied by numerous investigators and is comparatively easily determined by investigating the change in the specific gravity. It is, however, not possible to calculate or to reach any definite conclusions in regard to the dimensional changes from the change in volume during heat treatment because the changes in dimension are not equal in each direction. Changes in dimension during heat treatment are dependent upon numerous factors other than the structural changes, as for example, the shape of the object, conditions during heating and cooling, etc.

The change in dimension during heat treatment has not been very thoroughly studied in spite of its great practical importance; and it is the purpose of the present investigation to supplement previous studies with data relative to dimensional changes occurring during quenching and tempering of ordinary grades and shapes of tool steels. This investigation has been limited to include only the changes in dimensions occurring when the steel is being heated and cooled properly and, therefore, does not include the pronounced dimension changes, such as warpage, which might occur as a result of improper heating and cooling.

CAUSES FOR DIMENSION CHANGE DURING HEAT TREATMENT

The opinion prevalent heretofore concerning the change of dimensions during heat treatment, and which has been advocated by F. Berger (1)¹ among other investigators, has been that the object has a tendency to change its shape so as to assume a form approaching that of a sphere. There are exceptions to this rule as will be described later. In order to explain this phenomenon it will first be necessary to make an analysis of the various factors which influence dimension change during the hardening; and then follow with a study of the result of experiments.

When an object is deformed it is necessary that it has been subject to stresses of some kind. During the heat treatment of steel, stresses usually develop due to the following causes:

1. Nonuniform temperature distribution, which causes heat stresses.
2. A change in volume, which occurs during the structural changes in the steel and causes transformation stresses.

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

(Future reference will be made to transformation stresses as the stresses occurring during the transformation of austenite to martensite.)

DEFORMATION CAUSED BY HEAT STRESSES

If a material is heated or cooled, heat stresses will always develop because uneven temperature distribution can never be entirely avoided. Permanent deformation, however, occurs only when these stresses become strong enough to exceed the elastic limit of the material. For that reason, there will be no dimension changes on account of heat stresses in an object that is heated to and cooled from a relatively low temperature, partly on account of the fact that the heat stress then usually becomes relatively negligible, partly because the elastic limit of the material is still relatively high. This has been confirmed by Portevin and Sourdillon (2) who found no appreciable change of dimension after cooling steel cylinders 25 millimeters round by 145 millimeters (0.984 by 5.708 inches) from various temperatures below approximately 500 degrees Cent. (930 degrees Fahr.).

At higher temperatures the dimension changes rapidly increase, partly due to increased heat stresses, partly because the elastic limit of the material is decreased rapidly with increasing temperature. The greater the temperature differences in a piece of steel, the larger will be the heat stresses. The more rapidly the object is heated and cooled, the greater will be the deformation on account of heat stresses. The lower the heat conductivity of the material, the larger the temperature differences and deformation. Shapes, which favor the occurrence of large temperature differences, will cause larger deformation. A layer of scale, unevenly distributed over the surface, might also increase the temperature differences when hardening an object, and in this way cause larger deformation. The higher the elastic limit of the material at elevated temperatures, the smaller will be the deformation.

The following explanation is given as to how an object is deformed on account of heat stresses, which reasoning is in agreement with that of F. Berger (1), D. Hattori (3), and others:

When a piece of steel is cooled from a high temperature, the surface at first shrinks more rapidly than the center. The center will, therefore, be subjected to a pressure from all sides. We might

compare the object with a container, which is under pressure. Such a container tends to take such a shape that the volume will be as large as possible relative to the surface, that is, to assume spherical shape. If it is a question of a cube, the sides will tend to bulge. With a long cylinder, the length being longer than the diameter, the length will decrease and the diameter increase. With a short cylinder, the length being shorter than the diameter, the tendency will be the opposite. As a result of experiments carried out by D. Hattori (3), Portevin and Sourdillon (2), and also from this present investigation, it is evident that the change of dimensions really occur, according to the above theory, provided purely heat stresses occur and, therefore, no stresses are induced on account of transformations in the steel. We can, therefore, expect such a change in dimensions for all steels, which are cooled under the transformation temperature A_{c1} and for all steels which lack transformation point, such as the austenitic and purely ferritic steels.

It should be noted, however, that when the steel has been cooled down sufficiently, there will be a reversal of the movement described above. After a certain temperature has been reached, the center begins to cool faster than the surface and it tends to occupy a smaller volume than the surface. This reverse movement occurs at a low temperature of about 150 degrees Cent. (300 degrees Fahr.), according to H. Bühler and Buchholtz (4). The material at this low temperature has such little plasticity that only slight changes of dimensions can occur. The character of the change of dimensions, therefore, will be determined by the condition of stress present during the early part of the cooling.

DEFORMATION CAUSED BY TRANSFORMATION STRESSES

When hardening a piece of steel, there will be at first a change of dimension caused by heat stresses as per the above discussion. When the martensitic point has been reached, that is, at a temperature below approximately 300 degrees Cent. (570 degrees Fahr.) there will be a sudden expansion of the surface on account of the martensite having larger specific volume than the austenite from which it has been formed. The center will, therefore, be subjected to a tension from all sides and we can, therefore, consider that the piece of steel corresponds to a container in which there is vacuum. Such a container will change its shape opposite to that of a sphere,

i.e., an entirely contrary condition from that caused by heat stresses. On a cube, therefore, the sides will bulge in; on a long cylinder the diameter will decrease and the length will increase, and so forth.

When the core has been cooled sufficiently so that the formation of martensite also occurs in the center, a deformation should occur which is opposite to that just mentioned. Because of the surface having a martensitic structure with high elastic limit, there can only occur a relatively negligible equalization of the deformation.

We may formulate the theory that the lower the temperature of the martensitic transformation, the less deformation can be expected on account of transformation stresses because the resistance of the material against deformation is increased at the lower temperature. It is, however, not entirely certain that the deformation will be smaller. Since, at a lower temperature, the difference between the specific volume of the austenite and martensite will be greater and, therefore, the transformation stresses might be larger with the lower transformation temperatures. Compare H. Bühler and E. Scheil (5).

Using the theory given above, it can be explained why the change of dimensions occurring during the formation of martensite will not be equal in all directions, but that certain dimensions will increase more and others less and, also, that the change of dimension will be opposite to that caused by heat stresses.

If there are certain constituents in the structure, which do not participate in the transformation, for instance, the remaining austenite, carbide in hypereutectoid steels, or ferrite, these should decrease the deformation on account of transformation stresses, as only a certain part of the structure will change its volume during the quenching.

DEFORMATION DUE TO BOTH HEAT AND TRANSFORMATION STRESSES

When hardening a piece of steel, the deformation is probably never caused only by heat stresses or transformation stresses, but usually by both these factors. In order that steel may harden, a certain cooling velocity is necessary and, therefore, there are always certain heat stresses created during the quenching. The stresses which will determine the character of the deformation depend upon numerous factors. It is, however, to be expected that with air hardening steels, where the cooling velocity is relatively small, the defor-

mation will mainly be determined by the transformation stresses.

In order to further illustrate how heat stresses and transformation stresses influence the deformation, a cube of soft iron with a carbon content of 0.04 per cent, 20-millimeter side (0.787 inch), (soft iron 6, Table I) was cooled fifty times from 800 degrees Cent. (1470 degrees Fahr.), i.e., from below the transformation point Ac_3 . In this sample only heat stresses could have been present. Another



Fig. 1—Sections of Cubes of Air Hardening Steel No. 5, Table I, and Soft Iron No. 6, Table I, Heated and Cooled Fifty Times from 1000 and 800 Degrees Cent. Respectively.

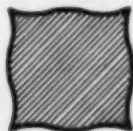


Fig. 2—Section of a Tool Steel Which Did Not Harden Through After Repeated Heating and Cooling.

similar sample, but of an air hardening grade (steel 5, Table I) was cooled fifty times in air from 1000 degrees Cent. (1830 degrees Fahr.). In this test transformation stresses should have been predominant. The appearance of the samples after these treatments is evident from Figs. 1 and 2, showing a section through the center of the samples parallel with two surfaces.

As may be seen from Fig. 1, the deformation has occurred in accordance with the theories previously given. The sample piece of the soft iron has evidently commenced to take a spherical shape on account of the heat stresses. With the high carbon, high chromium steel the deformation is just the opposite because the transformation stresses were dominating.

The behavior of steels, that will not harden entirely through, should now be considered. With such steels there occurs, during quenching, first a deformation on account of heat stresses, which can be expected to be relatively pronounced because such steels gen-

erally have to be cooled rapidly in order to take satisfactory hardening. When the surface hardens it expands, and deformation on account of transformation stresses will occur. When the core changes from austenite to troostite, sorbite or pearlite, it expands, which counteracts the deformation on account of the transformation stresses. The final result should be a deformation depending on both the heat and transformation stresses. W. Haufe (6) has obtained the deformation indicated in Fig. 2 when repeatedly hardening samples of tool steels which do not harden through. This deformation can be considered as being caused by a combination of heat and transformation stresses.

EXPERIMENTAL MATERIAL

The following types of steel were selected for this investigation because they are normally used for dies, taps and other tools where the dimension change during the heat treatment is especially important.

- No. 1—Very pure unalloyed carbon steel (carbon content 1.00 per cent) with wide hardening range, relatively shallow depth of hardness penetration, and water hardening.
- No. 2—Steel of the same analysis as No. 1, with not so wide a hardening range and greater depth of hardness penetration, water hardening.
- No. 3—Tungsten steel suitable for twist drills and similar tools, water hardening.
- No. 4—Chromium-tungsten-vanadium oil hardening, so-called "nondeforming" steel.
- No. 5—High carbon, high chromium air hardening "nondeforming" steel with molybdenum and vanadium additions.

In order to investigate the deformation occurring when only heat stresses are present, the following sample was also included in the test—

- No. 6—Very soft and pure iron manufactured by the Flodin process.

The analyses, hardening ranges, and depths of hardness penetration of the steels used in the investigation are shown in Table I. The fracture grain size is measured by the Swedish fracture grain size standard. The depth of hardness penetration has been deter-

mined according to the standard procedure used by the Uddeholms A/B, i. e., by measuring the depth of hardness penetration in millimeters from the fracture of a sample piece 20 millimeters round x 100 millimeters long (0.787 x 3.937 inches) which has been hardened and afterwards broken at the middle.

SIZE AND SHAPE OF SAMPLES

The following samples were selected keeping in mind the dimensions of some common tools, such as punches, gages, drills, broaches, threading chasers, reamers, etc.

Cylinder diameter		Height	
5 millimeter round	(0.196 inch)	15 millimeter	(0.590 inch)
5 millimeter round	(0.196 inch)	50 millimeter	(1.968 inch)
25 millimeter round	(0.984 inch)	145 millimeter	(5.708 inch)
50 millimeter round	(1.968 inch)	10 millimeter	(0.393 inch)
50 millimeter round	(1.968 inch)	50 millimeter	(1.968 inch)

Ring O. D.	I. D.	Height
15 millimeter rd. (0.590 in.)	5 millimeter rd. (0.196 in.)	5 millimeter (0.196 in.)
15 millimeter rd. (0.590 in.)	5 millimeter rd. (0.196 in.)	10 millimeter (0.393 in.)
75 millimeter rd. (2.952 in.)	25 millimeter rd. (0.984 in.)	25 millimeter (0.984 in.)
75 millimeter rd. (2.952 in.)	25 millimeter rd. (0.984 in.)	50 millimeter (1.968 in.)

PREPARATION AND HEAT TREATMENT OF SAMPLES

The material for the samples was available in the as-rolled and annealed condition and was selected from 7, 18, 28 and 85 millimeter sizes (0.275, 0.708, 1.102 and 3.345 inch). The samples were turned down with the greatest possible accuracy from the most suitable dimension and were polished with emery cloth.

In order to determine the most suitable heat treating temperature, experiments were carried out with 5, 25 and 50 millimeter round samples (0.196, 0.984 and 1.968 inch). These were quenched from various temperatures and then fracture grain size, hardness penetration and Rockwell hardness were determined. As a result of these tests, a suitable quenching temperature of each steel and sample size was selected for the principal investigation, and a temperature of 30 degrees Cent. (55 degrees Fahr.) higher was also chosen to study its effect on the samples.

The samples were preheated, packed in charcoal breeze and were finally heated in a salt bath. The tongs used during the quenching were made up especially to give the smallest possible con-

Table II

Steel	Sample Dimension				Sample Dimension				Sample Dimension			
	5 rd. x 15 mm., 5 rd. x 50 mm.		15 rd. x 5 rd. x 10 mm.		25 rd. x 25 rd. x 145 mm.		50 rd. x 25 rd. x 50 mm.		75 rd. x 25 rd. x 50 mm.		100 rd. x 25 rd. x 50 mm.	
	Hardening Temp.	Time for heating through minutes	Hardening Temp.	Time for heating through minutes	Hardening Temp.	Time for heating through minutes	Hardening Temp.	Time for heating through minutes	Hardening Temp.	Time for heating through minutes	Hardening Temp.	Time for heating through minutes
1	765° C. (1410° F.)	5	770° C. (1420° F.)	15	780° C. (1435° F.)	25	790° C. (1455° F.)	40	790° C. (1455° F.)	40	790° C. (1455° F.)	60
	795° C. (1465° F.)	5	800° C. (1470° F.)	15	810° C. (1490° F.)	25	820° C. (1510° F.)	40	820° C. (1510° F.)	40	820° C. (1510° F.)	60
2	760° C. (1400° F.)	5	765° C. (1410° F.)	15	775° C. (1425° F.)	25	785° C. (1445° F.)	40	785° C. (1445° F.)	40	785° C. (1445° F.)	60
	790° C. (1455° F.)	5	795° C. (1465° F.)	15	805° C. (1480° F.)	25	815° C. (1500° F.)	40	815° C. (1500° F.)	40	815° C. (1500° F.)	60
3	770° C. (1420° F.)	5	775° C. (1425° F.)	15	785° C. (1445° F.)	25	795° C. (1465° F.)	40	795° C. (1465° F.)	40	795° C. (1465° F.)	60
	800° C. (1470° F.)	5	805° C. (1480° F.)	15	815° C. (1500° F.)	25	825° C. (1515° F.)	40	825° C. (1515° F.)	40	825° C. (1515° F.)	60
4	780° C. (1435° F.)	5	785° C. (1445° F.)	15	805° C. (1480° F.)	25	815° C. (1500° F.)	40	815° C. (1500° F.)	40	815° C. (1500° F.)	60
	810° C. (1490° F.)	5	815° C. (1500° F.)	15	835° C. (1535° F.)	25	845° C. (1555° F.)	40	845° C. (1555° F.)	40	845° C. (1555° F.)	60
5	995° C. (1825° F.)	5	1005° C. (1840° F.)	15	1010° C. (1850° F.)	25	1020° C. (1870° F.)	40	1020° C. (1870° F.)	40	1020° C. (1870° F.)	60
	1025° C. (1875° F.)	5	1035° C. (1895° F.)	15	1040° C. (1905° F.)	25	1050° C. (1920° F.)	40	1050° C. (1920° F.)	40	1050° C. (1920° F.)	60

Steels 1, 2 and 3 Quenched in Brine.

Steel 4 Quenched in Oil.

Steel 5 Quenched in Air.

Steels 1, 2, 3 and 4 Preheated 1 Hour at 700° C. (1290° F.).

Steel 5 Preheated at 750° C. (1380° F.).

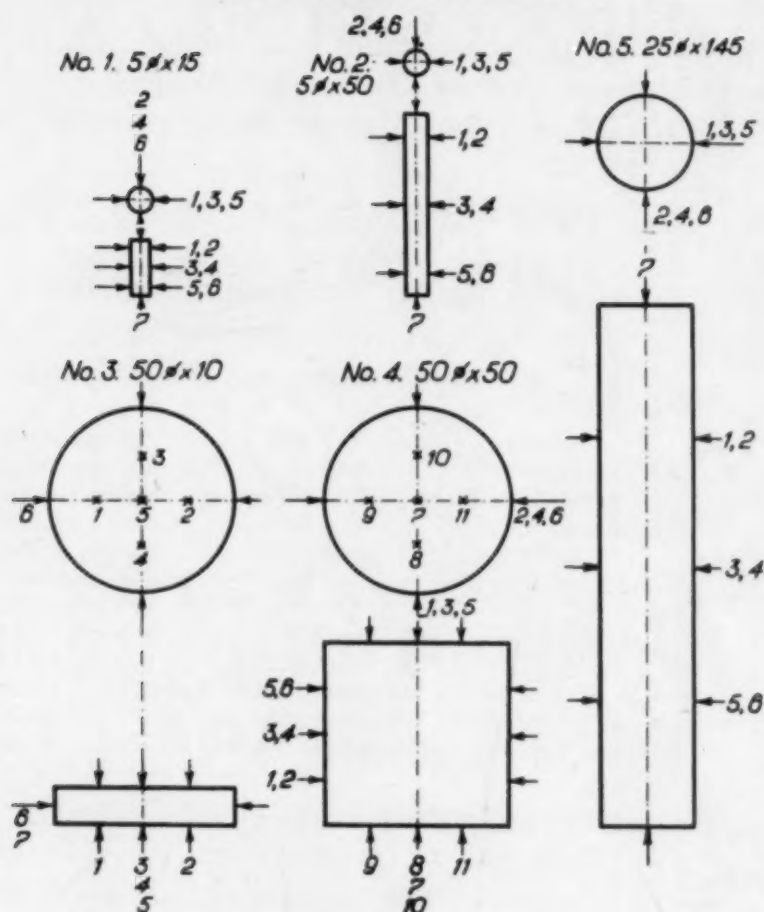
Measuring Points on Cylinders.

Fig. 3—Diagram Showing the Location of the Measuring Points on Cylinders.

tact surface between the jaws and the sample. The samples of the dimensions 5 millimeter round x 15 millimeter; 5 millimeter round x 50 millimeter; 25 millimeter round x 145 millimeter; 50 millimeter round x 50 millimeter, and 75 millimeter round x 25 millimeter round x 50 millimeter were immersed in the quenching medium with the axes vertical. The remaining samples were immersed with the axes horizontal. The quenching temperatures, cooling times, etc., are shown in Table II. All samples of the soft iron, No. 6, were quenched in brine from 800 degrees Cent. (1470 degrees Fahr.).

In order to determine the amount of scale formed during the heating operation, the samples were weighed before and after the quench, and the thickness of the oxidized material was thus obtained. These figures varied between 0.001 to 0.10 millimeter (0.000039 to 0.000393 inch), depending upon kind of steel and quenching tem-

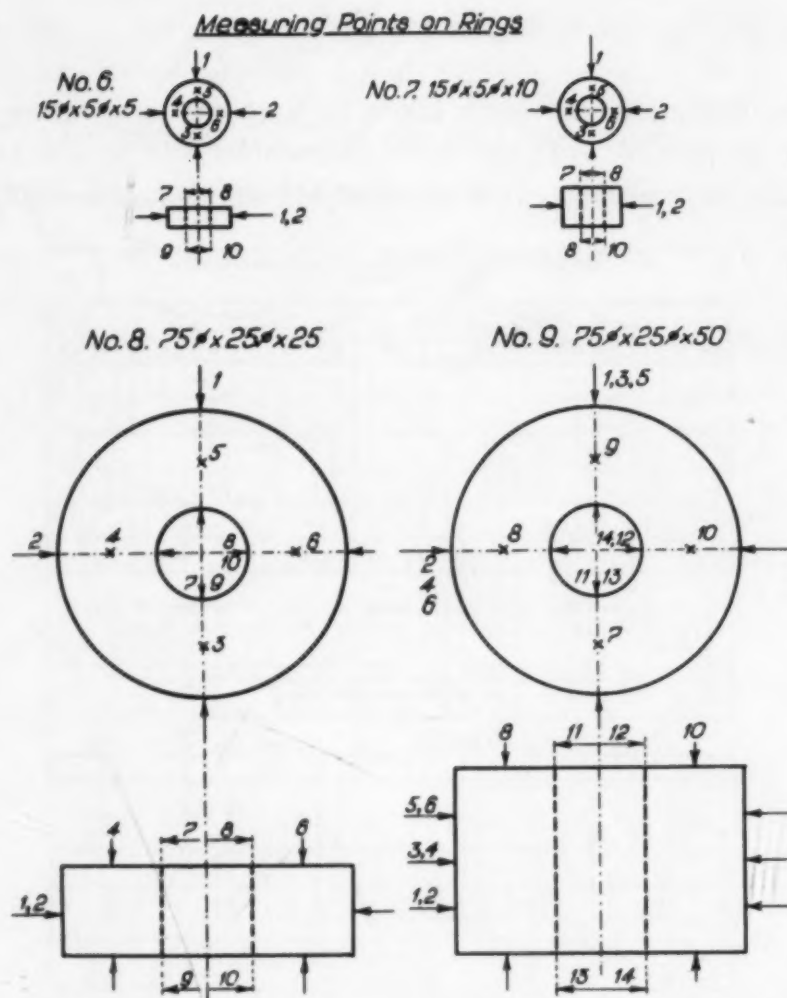


Fig. 4—Diagram Showing the Location of the Measuring Points on Rings.

perature. Correction for this loss has been made in the figures shown for the dimension change. All samples were checked for Rockwell hardness in order to determine if the quenching operation had been carried out satisfactorily.

Tempering of the samples was done with slow heating and cooling with a holding time of 30 minutes. The tempering furnace was equipped with air circulation in order to obtain uniform temperature distribution. The following drawing temperatures were used:

Steel	Tempering Temperature					
1—2—3	100° C. (210° F.)	150° C. (300° F.)	200° C. (390° F.)	250° C. (480° F.)	300° C. (570° F.)	350° C. (660° F.)
4	100° C. (210° F.)	150° C. (300° F.)	200° C. (390° F.)	250° C. (480° F.)	300° C. (570° F.)	350° C. (660° F.)
5	100° C. (210° F.)	150° C. (300° F.)	200° C. (390° F.)	250° C. (480° F.)	300° C. (570° F.)	350° C. (660° F.)
	400° C. (750° F.)	450° C. (840° F.)	500° C. (930° F.)	550° C. (1020° F.)	600° C. (1110° F.)	400° C. (750° F.)

MEASURING METHOD

The measurements were taken at a room temperature kept as constant as possible with the same micrometer and by the same person. The micrometers were checked before each measuring opera-

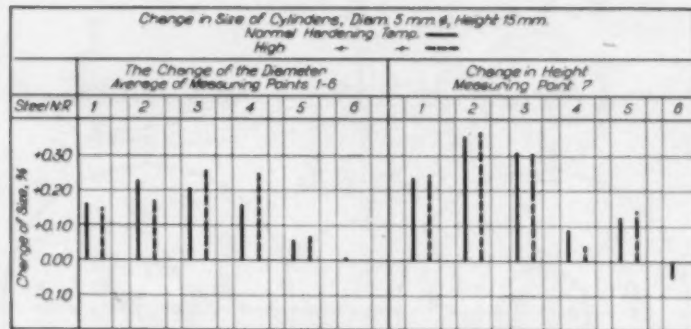


Fig. 5—Graphic Illustration of Size Change of Cylinders 5 Millimeters Round and 15 Millimeters Long.

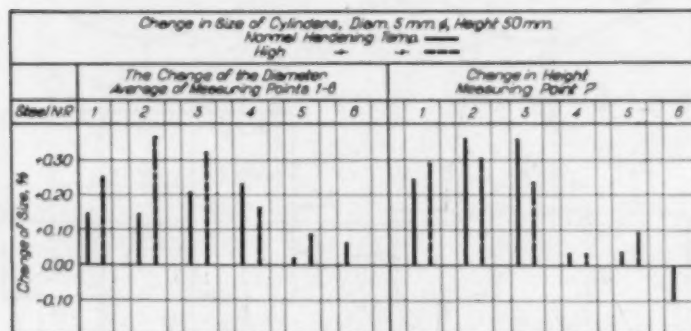


Fig. 6—Graphic Illustration of Size Change of Cylinders 5 Millimeters Round and 50 Millimeters Long.

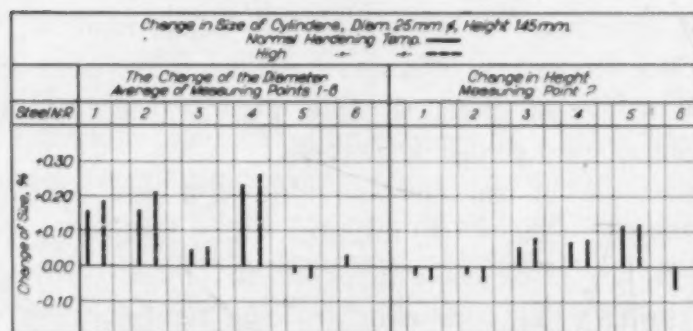


Fig. 7—Graphic Illustration of Size Change of Cylinders 25 Millimeters Round and 145 Millimeters Long.

tion with Johansson precision gages. The measurements were taken before and after the quenching and after the various drawing temperatures in identically the same spots, which were indicated by

punch marks on the samples. The locations of the various measuring points are shown in Figs. 3 and 4.

DEFORMATION CAUSED BY QUENCHING

In Figs. 5 to 13 the result from the hardening tests with the

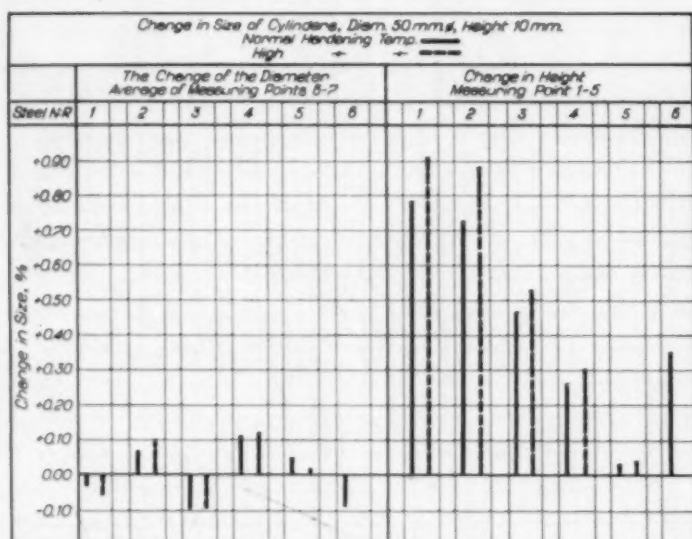


Fig. 8—Graphic Illustration of Size Change of Cylinders 50 Millimeters Round and 10 Millimeters Long.

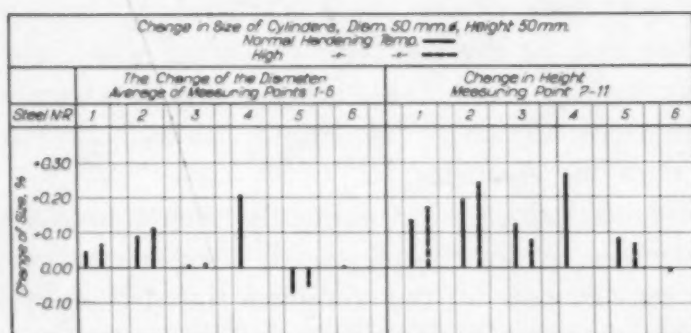


Fig. 9—Graphic Illustration of Size Change of Cylinders 50 Millimeters Round and 50 Millimeters Long.

several steels is graphically shown. The dimension change has been expressed in per cent, plus if the dimension has increased during quenching, and minus if it has been reduced.

CYLINDERS, DIAMETER 5 MM. ϕ , HEIGHT 15 MM.

From Fig. 5 it is evident that with steels Nos. 1, 2 and 3 the length and the diameter have been increased considerably in the 5-millimeter round x 15-millimeter cylinders. The increase of vol-

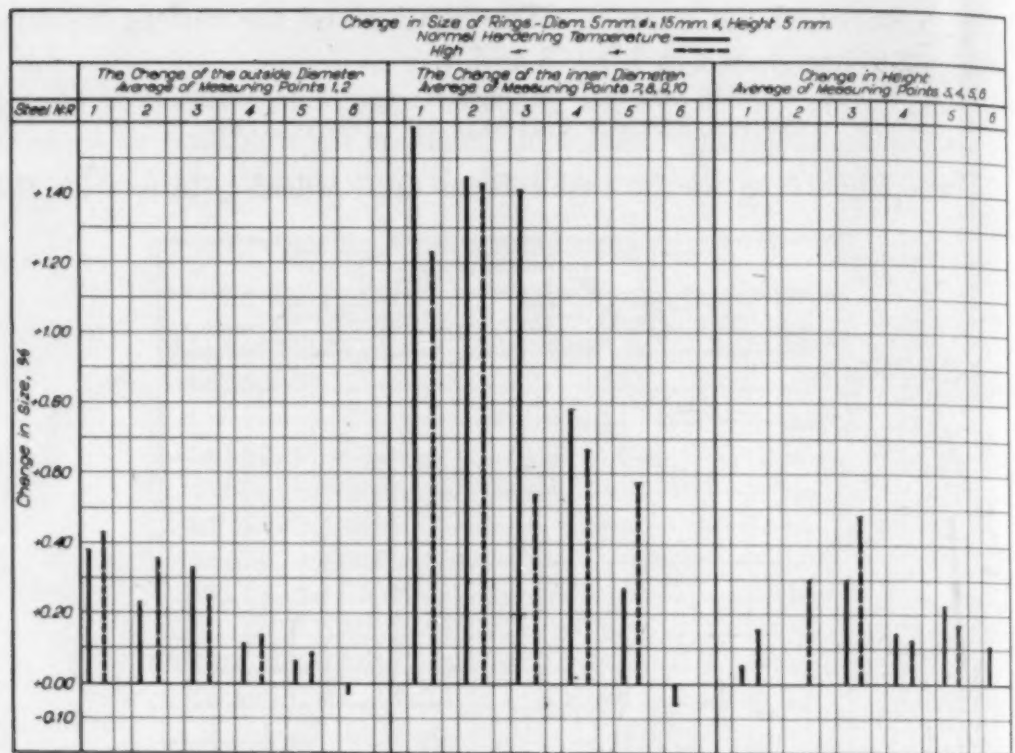


Fig. 10—Graphic Illustration of Size Change of Rings 5 and 15 Millimeters in Diameter and 5 Millimeters Long.

ume during quenching of the steels is, therefore, large in this case, which is explained by the fact that the small dimension has hardened entirely through, and thus showing that martensite is formed in the entire cross section. The length has increased to a greater percentage than the diameter.

The dimension change has evidently been caused by the transformation stresses, as per the previously mentioned theory. This is explainable on account of the complete formation of martensite. With only heat stresses, the dimension change occurs in the opposite direction, as is evident from the test with the sample of soft iron (No. 6).

With steel No. 4 the dimension change, and, therefore, also the volume change, is smaller. This is explained by the fact that martensite has not been formed exclusively and that a certain amount of austenite is present which has smaller specific volume than martensite. That austenite is being created in not an inconsiderable amount, when quenching this quality from the temperature in question, is evident by the fact that a relatively large increase of volume is obtained by tempering at about 250 degrees Cent. (480 degrees

With steel No. 5, the volume change is very small, which is explained partly by the fact that the amount of the remaining austenite is relatively large, and partly that the carbide content, which does not participate in the transformation and, thus, does not influence the volume change, is high. According to B. Kalling (7) the carbide in a chromium steel with about 13 per cent chromium has a carbon content of approximately 5.35 per cent and, according to V. N. Krivobok (8), the austenite in such a steel has at saturation at a temperature of 1000 degrees Cent. (1830 degrees Fahr.) a carbon content of about 0.50 per cent. If we calculate with these values, it is evident that steel No. 5 with a carbon content of 1.64 per cent, at a hardening temperature of about 1000 degrees Cent. (1830 degrees Fahr.), has approximately 23.5 per cent of its structure in form of carbides, which do not participate in the transformation of the steel.

Steel No. 5 has increased in length more than the diameter. The change in dimension has, therefore, been determined by the transformation stresses, which is quite natural as the heat stresses ought to be quite small on account of the air cooling of the sample.

When hardening from the higher temperature about the same dimension change has been obtained as at the lower temperature for all steels, with the exception of steel No. 4.

CYLINDERS, DIAMETER 5 MM. ϕ , HEIGHT 50 MM.

As is evident from Fig. 6, the cylinders, dimension 5 millimeter round x 50 millimeter, behave in much the same way as the shorter cylinders, dimension 5 millimeter round x 15 millimeter, at normal hardening temperature. At the higher hardening temperature for steels Nos. 1, 2, 3, and 5, the dimension change has been caused more by the heat stresses.

CYLINDERS, DIAMETER 25 MM. ϕ , HEIGHT 145 MM.

With steels Nos. 1 and 2, the diameter has increased and the length has decreased, as is shown in Fig. 7. This is contrary to the action of the small cylinders made from the same steels and may be explained by the fact that the large cylinders have not hardened through and that the hardness penetration is relatively small and, therefore, the dimension change is determined by the heat stresses in the unhardened portion. That the heat stresses change the sam-

ple piece this way is evident from the experiment with the soft iron (No. 6).

With steel No. 3, which also has shallow hardness penetration, we might expect similar dimensional changes. With this quality, however, approximately the same percentage change in diameter and length has been obtained. This may be explained by the fact that this material, on account of its tungsten content, has higher tensile strength at elevated temperatures and, therefore, the heat stresses have not had such a pronounced effect.

The steels Nos. 4 and 5 behave in much the same way in the 25-millimeter round x 145-millimeter cylinders as in the small cylinders. Steel No. 4 has thus changed its form on account of heat stresses, and No. 5 on account of transformation stresses. This is, however, to be expected as these steels harden all the way through even in large dimensions and, therefore, no change in principle has occurred as compared to the small cylinders.

CYLINDERS, DIAMETER 50 MM. ϕ , HEIGHT 10 MM.

All steels, with the exception of No. 5, as is shown in Fig. 8, show a pronounced increase in the thickness while the diameter has changed very slightly. The change of form of these samples has evidently been caused by heat stresses. Compare these changes with the soft iron sample. On account of the shape of the sample, it thus seems that the heat stresses might cause an especially severe deformation. At the higher hardening temperature, the deformation has increased still more on account of larger heat stresses than at the normal hardening temperature.

With steel No. 5, the change in shape is very insignificant and occurs in the opposite direction at normal hardening temperature. The deformation, in this case, as well as with previous samples of steel No. 5, has been determined by the transformation stresses. From this test we can reach the conclusion that the change in dimension of steel No. 5 is caused almost exclusively by transformation stresses.

CYLINDERS, DIAMETER 50 MM. ϕ , HEIGHT 50 MM.

The heat stresses seem to cause only slight deformation on this size sample, as is evident from the experiment with the soft iron sample, Fig. 9. It is, therefore, to be expected that the transformation stresses will determine the deformation.

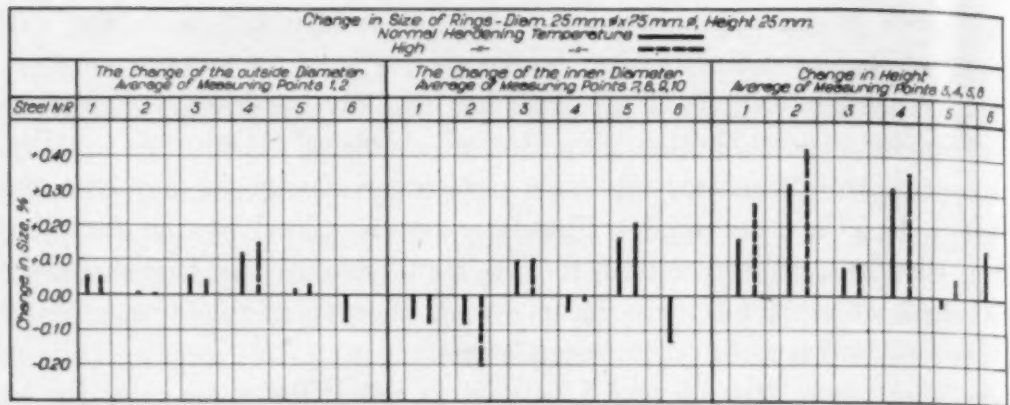


Fig. 12—Graphic Illustration of Size Change of Rings 25 and 75 Millimeters in Diameter and 25 Millimeters Long.

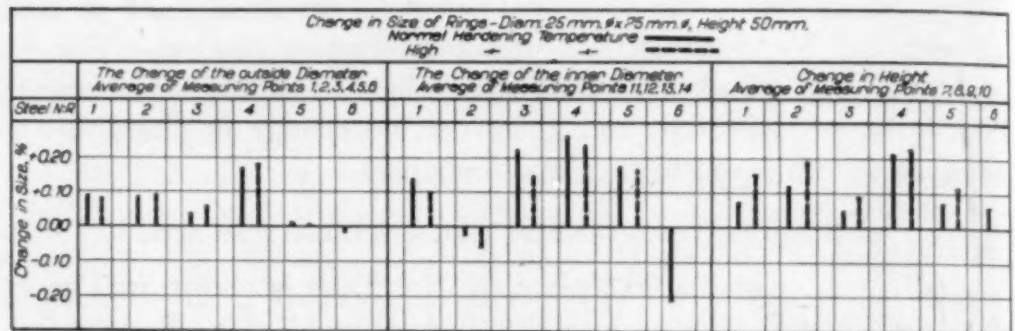


Fig. 13—Graphic Illustration of Size Change of Rings 25 and 75 Millimeters in Diameter and 50 Millimeters Long.

If we assume that steel No. 5, in accordance with its action in the samples previously examined, has changed due to transformation stresses, it is evident that the transformation stresses of this cylinder, as well as of the cylinders examined with the length larger than the diameter, increase the length more than the diameter. The length has increased more than the diameter, even with the other steels and, therefore, the transformation stresses would, in all instances, have determined the nature of the change form. Apparently, the heat stresses are of small importance, in view of the action of the soft iron (No. 6) sample, Fig. 9. The cylinders made from steels Nos. 1, 2 and 3 do not harden through and, on account of this, have changed less than steel No. 4.

RINGS, O.D. 15 MM. ϕ , I.D. 5 MM. ϕ , HEIGHT 5 AND 10 MM.

As may be seen from Figs. 10 and 11, the outside and inside diameters have decreased in the rings made from soft iron, which have only been subjected to heat stresses. This may be explained in the following manner: When cooling a ring, the outer surface

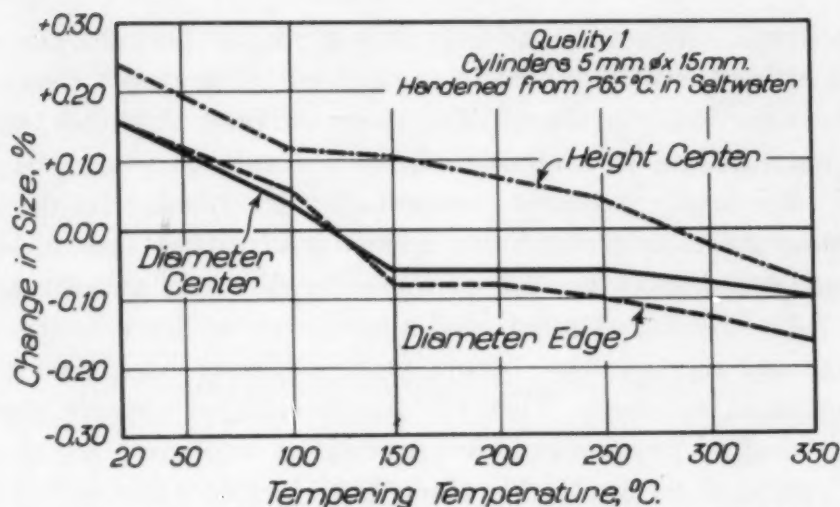


Fig. 14—Graphic Illustration of Size Change of Cylinders (Quality 1) 5 Millimeters in Diameter by 15 Millimeters, Hardened from 765 Degrees Cent. in Salt Water and Tempered.

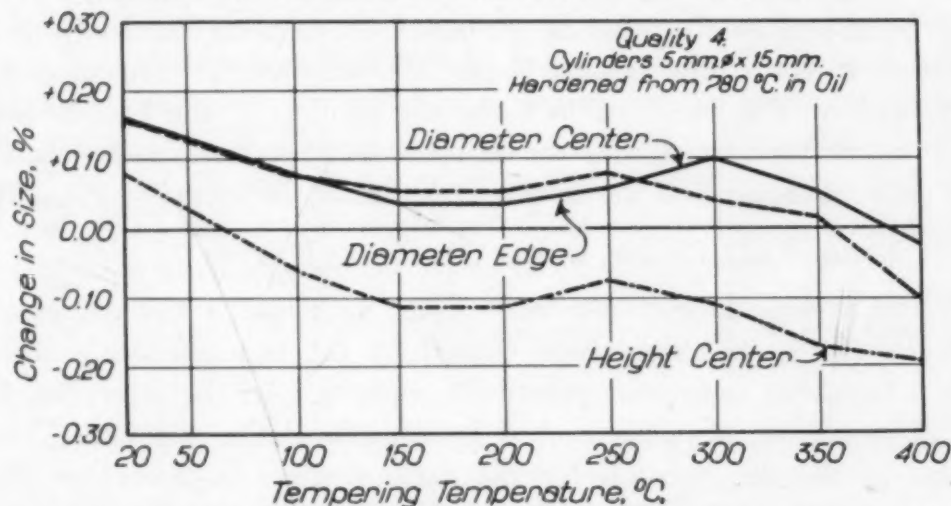


Fig. 15—Graphic Illustration of Size Change of Cylinders (Quality 4) 5 Millimeters in Diameter by 15 Millimeters, Hardened from 780 Degrees Cent. in Oil and Tempered.

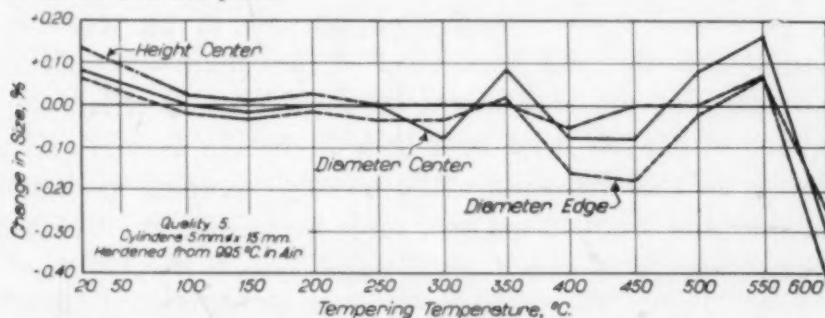


Fig. 16—Graphic Illustration of Size Change of Cylinders (Quality 5) 5 Millimeters in Diameter by 15 Millimeters, Hardened from 995 Degrees Cent. in Air and Tempered.

is being cooled more rapidly than the inner surface because the radiation of heat from the inside is not so rapid, especially when the

hole is small. When quenching in water, gas pockets are easily formed inside the hole, which to a great extent prevents removal of heat from the inner surface. The outer surface, therefore, shrinks faster than the inner surface and, as a result, the latter is compressed and finally becomes permanently deformed. In the rings of all the other steels, the inner diameter has, after the hardening, increased considerably more than the outer diameter and the height. As the heat stresses, as mentioned above, seek to decrease the inner diameter, any increase that occurs must evidently be caused by the transformation stresses. That the transformation stresses are able to affect such a pronounced deformation is explained by the fact that when the outer surface is being transformed from austenite to martensite and is expanded, the inner surface, on account of the slower cooling, is still plastic and can follow in the expansion. As may be noted, the increase in the inner diameter is especially pronounced in the thicker rings, length 10 millimeters. This may be explained by the fact that in those the cooling at the hole is less than in the thin rings, and the material is, therefore, more plastic and will be deformed more when the outer surface is expanded.

RINGS, O.D. 75 MM. ϕ , I.D. 25 MM. ϕ , HEIGHT 25 MM.

The inner diameter for steels Nos. 1, 2 and 4 has decreased and the change has thus been caused by the heat stresses. Steel No. 3 has been deformed practically equally. In the steel No. 5, the deformation has been caused by transformation stresses. The nature of the deformation for the various steels is, therefore, the same as for the cylinders, dimension 25 millimeter round x 145 millimeter, and the same explanation for the cause of the deformation as given for these tests should be applicable in the present case. It is quite natural that the same kind of deformation is obtained in these two cases, as the ring might be considered as a bar 25 millimeter square which has been bent in a circle like a ring, and as the hole is larger in the large rings, the cooling becomes more uniform on the inside and on the outside, and, therefore, the deformation will not be so severe as in the small rings.

RINGS, O.D. 75 MM. ϕ , I.D. 25 MM. ϕ , HEIGHT 50 MM.

It may be said generally that these rings with greater length have been deformed more than the preceding shorter rings on account of transformation stresses. The relation between the larger

rings of different length is, therefore, the same as for the small rings of different length, and the same explanation may be applied, i.e., that on account of the poorer cooling in the hole in the longer rings, the inner surface has higher temperature and is more plastic as the martensite transformation occurs in the outer surface. The inner surface is, therefore, more able to follow the expansion of the outer surface, and the deformation, on account of transformation stresses, will be greater.

COMPARATIVE DEFORMATION OF STEELS

Nos. 1, 2, 3, 4 AND 5

In order to obtain a measure of the tendency of various grades of steel to deform, the average has been obtained of the measured dimensional changes for one and the same quality, independent of whether the change has been plus or minus. The average deformation thus obtained probably could justly be considered as a rather accurate measure of the tendency to deform of the grade in question. As the depth of hardening has a great influence upon the deformation, the samples have been classified in two groups, those that have hardened all the way through, and those where the steels Nos. 1, 2 and 3 have not hardened through.

Quality	Average Deformation Per Cent	
	Small Dimensions	Large Dimensions
1	0.47	0.10
2	0.49	0.13
3	0.41	0.07
4	0.24	0.19
5	0.13	0.08

A discussion of the above classification follows:

Small Dimensions—Steels Nos. 1 and 2 have been deformed most and to approximately the same extent. Steel No. 3 is somewhat better, which may be explained by the fact that this grade on account of the tungsten content has higher tensile strength at elevated temperatures and is, therefore, not deformed so much.

It will be noted that steel No. 4 has deformed only half as much as the carbon steels. This may depend upon the following factors: Less heat stresses on account of oil quench, greater tensile strength at elevated temperatures and, finally, less change of volume on account of retained austenite in the structure.

Steel No. 5 shows only about one-quarter as much dimension change as the carbon steels and one-half as much as steel No. 4. This may be understood when one considers the small heat stresses

resulting from an air quench, the great amount of retained austenite, and by the fact that the structure of steel No. 5 consists, to a great extent, of carbides which do not participate in the volume change and, finally, by the higher tensile strength at elevated temperatures.

Large Dimensions—The different steels act entirely different in the larger size samples. Of the steels that do not harden through the entire section, steel No. 3 shows the least deformation, then steel No. 1, the poorest showing being made by steel No. 2—all of which depends on the depth of hardness penetration of the various steels, steel No. 3 having the shallowest and steel No. 2 the deepest penetration. Steel No. 4 shows much more deformation than steels Nos. 1, 2 and 3, which may be explained by the fact that the hardness penetration is so much deeper for this steel.

A very small deformation has been obtained with steel No. 5, notwithstanding a deep hardness penetration, and the change is practically the same as for steel No. 3, which is the best of all the steels in the larger size samples. The insignificant deformation of steel No. 5, as compared with steel No. 4, in spite of identically the same hardness penetration, may be explained in the same way as the difference between these two steels in the smaller samples, where steel No. 5 has also shown only about half as great change in dimension as steel No. 4.

THE INFLUENCE OF THE QUENCHING TEMPERATURE

It might be expected that an increase in quenching temperature would always increase the deformation, but this does not necessarily have to be the case. It is true, however, that the deformation on account of heat stresses should increase with raised quenching temperature, but this increase can be compensated for by the transformation stresses and the result might, therefore, be unchanged or decreased deformation.

If the average deformation is calculated for the several steels at high and low quenching temperature, and the change due to the raised quenching temperature is determined therefrom, the following result is obtained.

Quality	Change of average deformation quenched 30 degrees Cent. above normal quenching temperature	
	Per Cent	
1	Plus	11
2	Plus	23
3	Minus	12
4	Plus/Minus	0
5	Plus	9

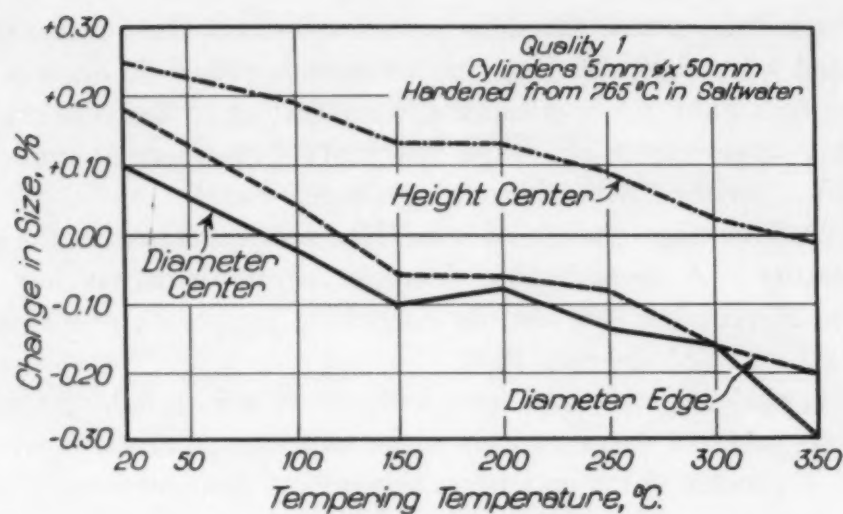


Fig. 17—Graphic Illustration of Size Change of Cylinders (Quality 1) 5 Millimeters in Diameter by 50 Millimeters, Hardened from 765 Degrees Cent. in Salt Water and Tempered.

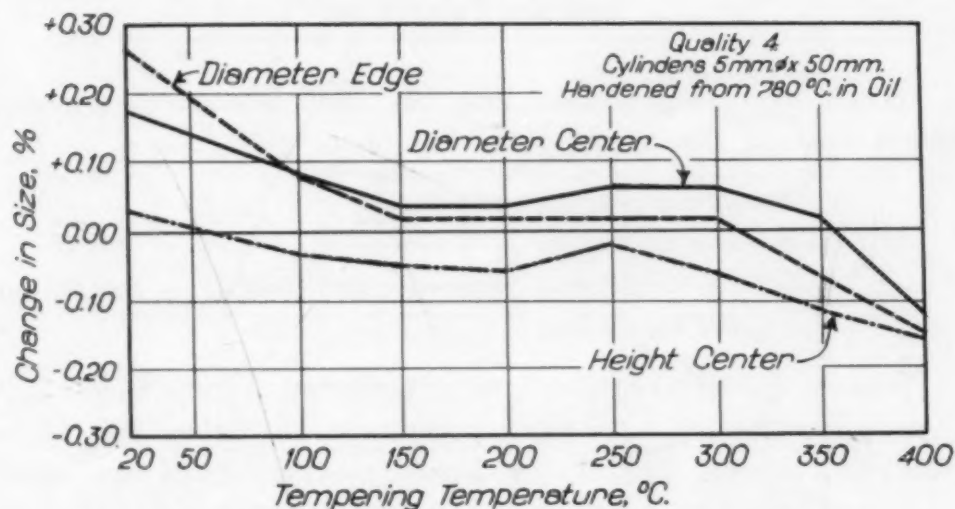


Fig. 18—Graphic Illustration of Size Change of Cylinders (Quality 4) 5 Millimeters in Diameter by 50 Millimeters, Hardened from 780 Degrees Cent. in Oil and Tempered.

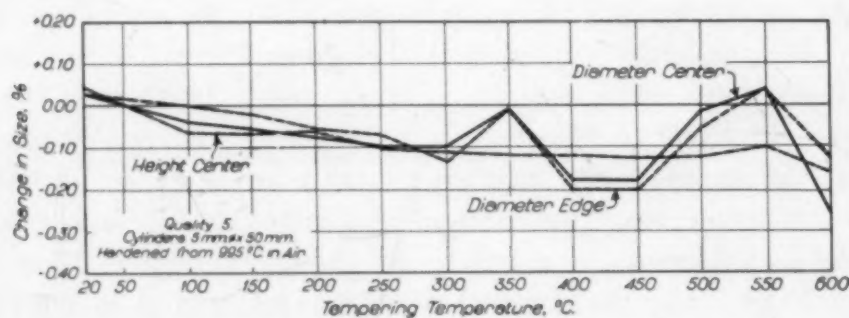


Fig. 19—Graphic Illustration of Size Change of Cylinders (Quality 5) 5 Millimeters in Diameter by 50 Millimeters, Hardened from 995 Degrees Cent. in Air and Tempered.

As will be noted, the deformation of steels Nos. 1, 2 and 5 is increased when the quenching temperature is raised 30 degrees Cent. (55 degrees Fahr.). It is especially interesting to see that steel No. 2 with poorer hardening range has moved considerably more than steel No. 1 with a wider hardening range.

The dimension change of steel No. 4 is the same at the raised temperature. A considerable decrease of deformation has taken place with steel No. 3 when the quenching temperature is raised 30 degrees Cent. (55 degrees Fahr.).

It is evident from these tests with the raised quenching temperature that, in most cases as it is to be expected, there is a tendency toward a greater deformation on account of heat stresses.

THE DIMENSIONAL AND STRUCTURAL TRANSFORMATIONS DURING TEMPERING

When tempering a steel that has been quenched, dimension changes may be due to the following factors:

- 1—Structural transformations
- 2—Release of strains
- 3—Heat strains on account of rapid heating or cooling during the tempering

When a quenched carbon steel is tempered, some changes occur in the martensite at a very low temperature, i.e., below 100 degrees Cent. (210 degrees Fahr.), and such transformations are accompanied by a reduction in volume—compare T. Berglund (13). Judging from the investigations of G. Hägg (9) it is possible that this contraction is connected with the precipitation of a carbide Fe_2C from the solid solution of C in alpha iron which the martensite represents. At approximately 160 to 170 degrees Cent. (320 to 340 degrees Fahr.) this decomposition is completed. At about 200 to 240 degrees Cent. (392 to 465 degrees Fahr.) austenite, which might be present, decomposes into martensite with an increase of volume. Above this temperature the martensite changes rapidly into troostite, resulting in a decrease in volume. At about 500 degrees Cent. (930 degrees Fahr.) the decomposition of the martensite is completed. These transformations might be moved toward higher temperatures with alloy steels.

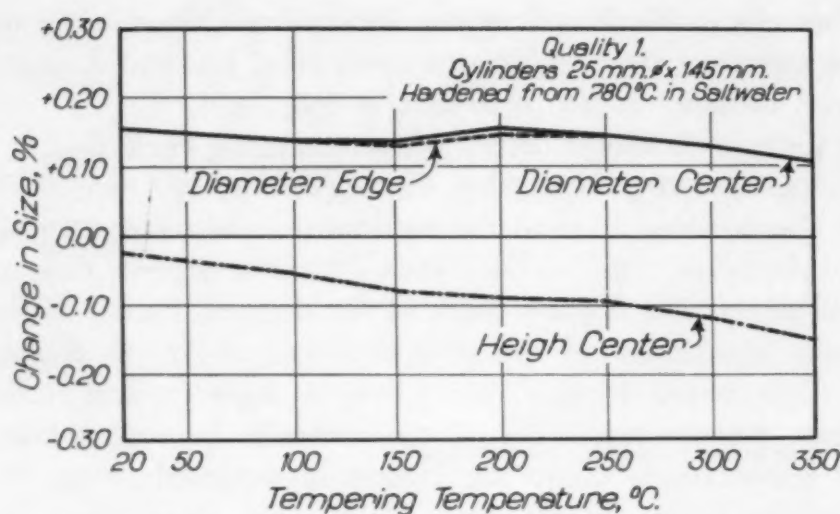


Fig. 20—Graphic Illustration of Size Change of Cylinders (Quality 1) 25 Millimeters in Diameter by 145 Millimeters, Hardened from 780 Degrees Cent. in Salt Water and Tempered.

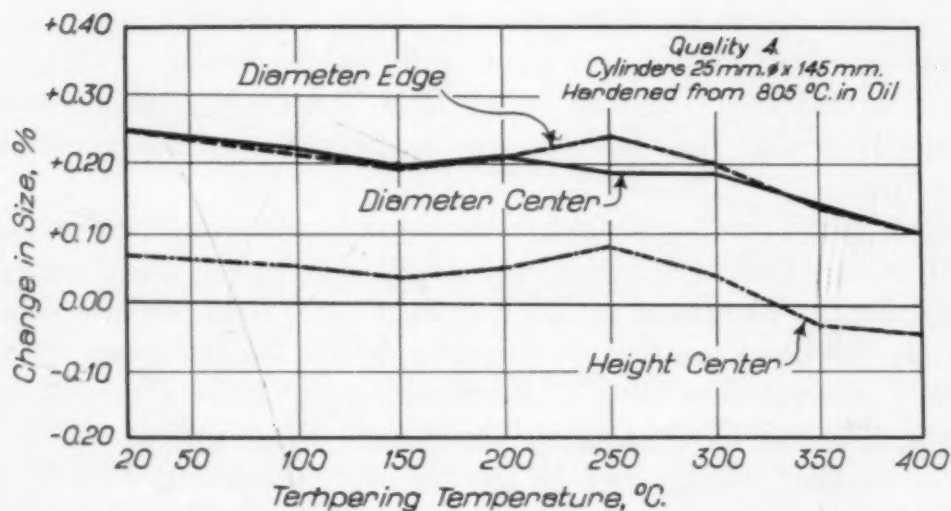


Fig. 21—Graphic Illustration of Size Change of Cylinders (Quality 4) 25 Millimeters in Diameter by 145 Millimeters, Hardened from 805 Degrees Cent. in Oil and Tempered.

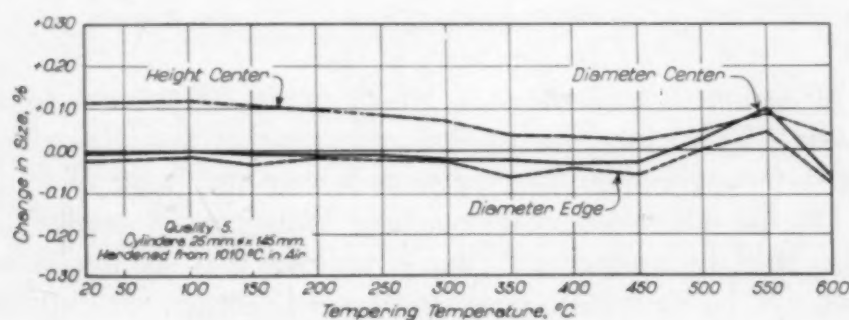


Fig. 22—Graphic Illustration of Size Change of Cylinders (Quality 5) 25 Millimeters in Diameter by 145 Millimeters, Hardened from 1010 Degrees Cent. in Air and Tempered.

Figs. 14 to 40 show how the dimensions change when tempering the quenched samples. Of the steels Nos. 1, 2 and 3, only No. 1 has been included as all three behave much alike.

Usually a shrinkage occurs when tempering steels Nos. 1, 2 and 3 to about 100 to 150 degrees Cent. (210 to 300 degrees Fahr.) due to the decomposition of the martensite. As the tempering temperature is raised from 150 degrees Cent. (300 degrees Fahr.) up to approximately 250 degrees Cent. (480 degrees Fahr.) there is a noticeable diminution of the shrinkage and at 200 to 250 degrees Cent. (390 to 480 degrees Fahr.) even a slight expansion, depending upon the decomposition of the austenite into martensite. At higher temperatures up to the highest investigated temperature of 350 degrees Cent. (660 degrees Fahr.) for these three steels, a shrinkage occurs as a result of the decomposition of martensite into troostite. The condition is similar with steel No. 4, but here the expansion, due to the decomposition of the austenite, is much more pronounced at 250 degrees Cent. (480 degrees Fahr.). This expansion will result in some cases in a dimension change proportionately greater than when it was in the quenched but not drawn condition.

A fairly gradual shrinkage occurs with steel No. 5 up to about 450 degrees Cent. (840 degrees Fahr.). Thereafter, there occurs a very distinct expansion at 550 degrees Cent. (1020 degrees Fahr.) due to the decomposition of austenite, which for this type of steel occurs at a considerably higher temperature than for the others investigated, and finally a rapid shrinkage. Even with steel No. 5, the dimension change may be greater after tempering than in the quenched condition, depending upon the decomposition of the austenite.

RELEASE OF STRAINS DURING TEMPERING

Those dimensional changes, which occur on account of transformation of structure during the tempering of samples which are hardened through, should be proportionately the same in all directions. If the dimension change during tempering is greater in one direction than in another, this would indicate that strains have been released. It is, therefore, possible to reach certain conclusions regarding the condition of strains after the quench from the size of the dimension change in different directions after tempering. Should

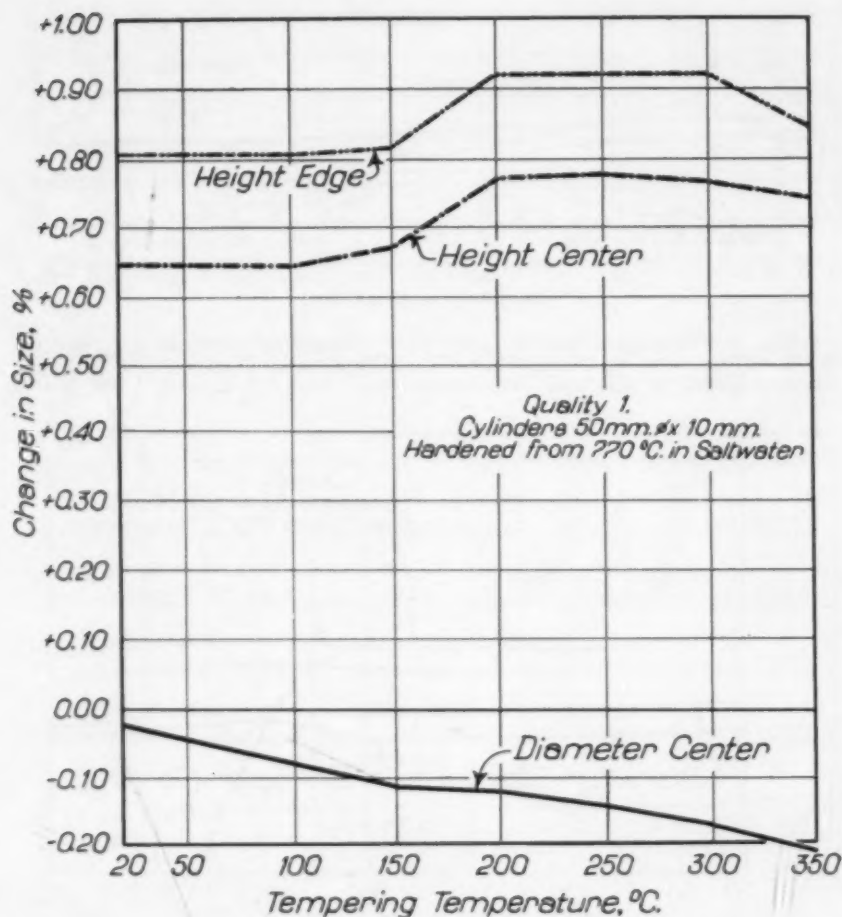


Fig. 23—Graphic Illustration of Size Change of Cylinders (Quality 1) 50 Millimeters in Diameter by 10 Millimeters, Hardened from 770 Degrees Cent. in Salt Water and Tempered.

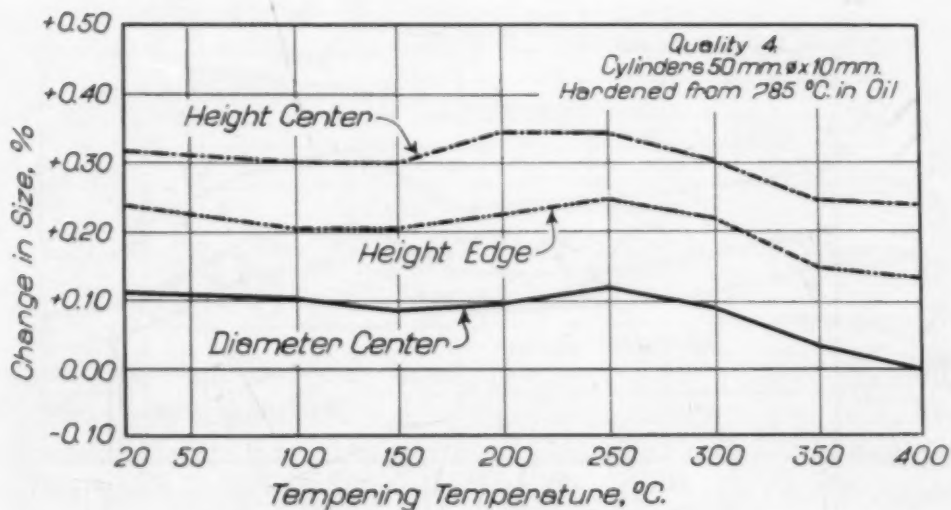


Fig. 24—Graphic Illustration of Size Change of Cylinders (Quality 4) 50 Millimeters in Diameter by 10 Millimeters, Hardened from 785 Degrees Cent. in Oil and Tempered.

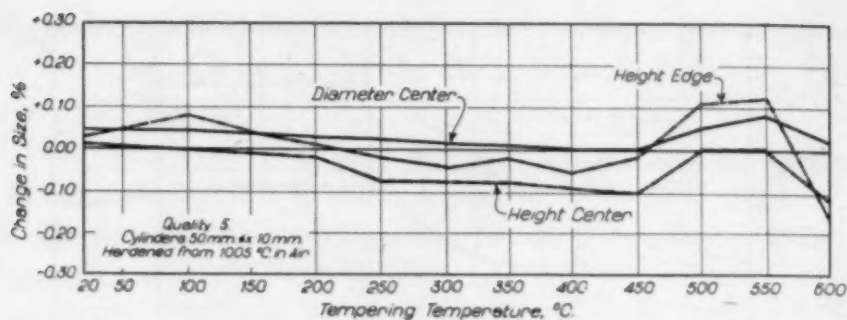


Fig. 25—Graphic Illustration of Size Change of Cylinders (Quality 5) 50 Millimeters in Diameter by 10 Millimeters, Hardened from 1005 Degrees Cent. in Air and Tempered.

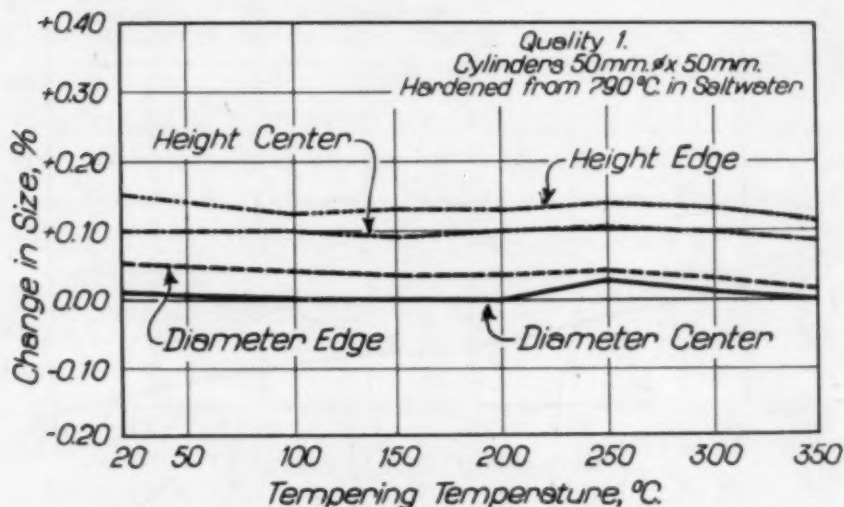


Fig. 26—Graphic Illustration of Size Change of Cylinders (Quality 1) 50 Millimeters in Diameter by 50 Millimeters, Hardened from 790 Degrees Cent. in Salt Water and Tempered.

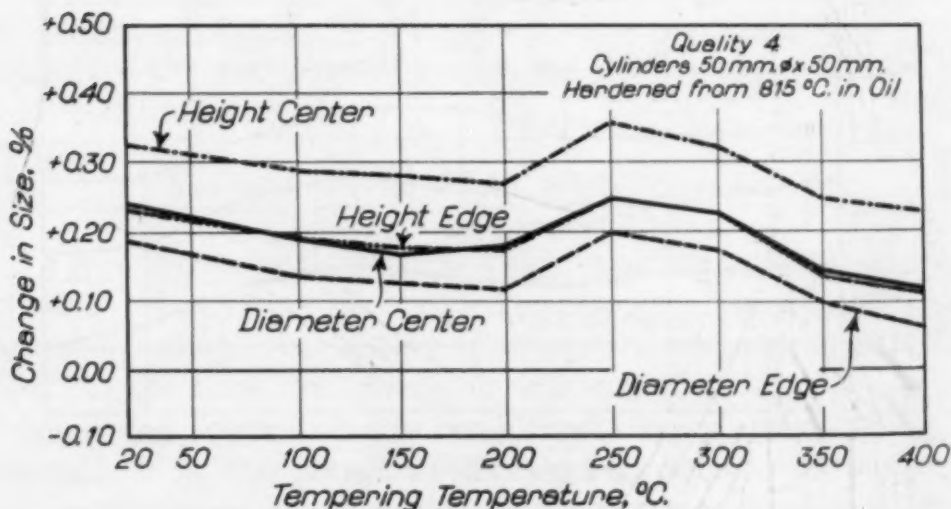


Fig. 27—Graphic Illustration of Size Change of Cylinders (Quality 4) 50 Millimeters in Diameter by 50 Millimeters, Hardened from 850 Degrees Cent. in Oil and Tempered.

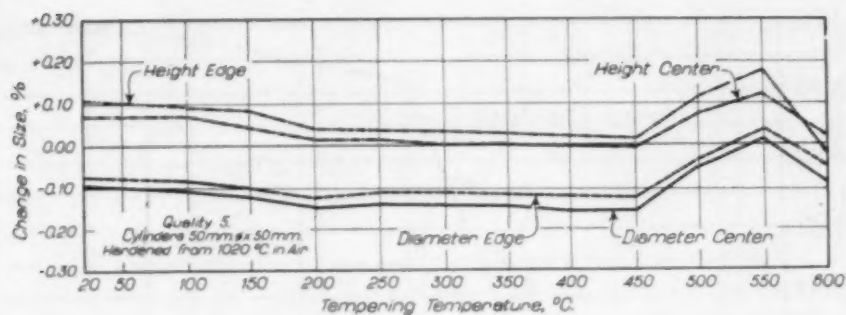


Fig. 28—Graphic Illustration of Size Change of Cylinders (Quality 5) 50 Millimeters in Diameter by 50 Millimeters, Hardened from 1020 Degrees Cent. in Air and Tempered.

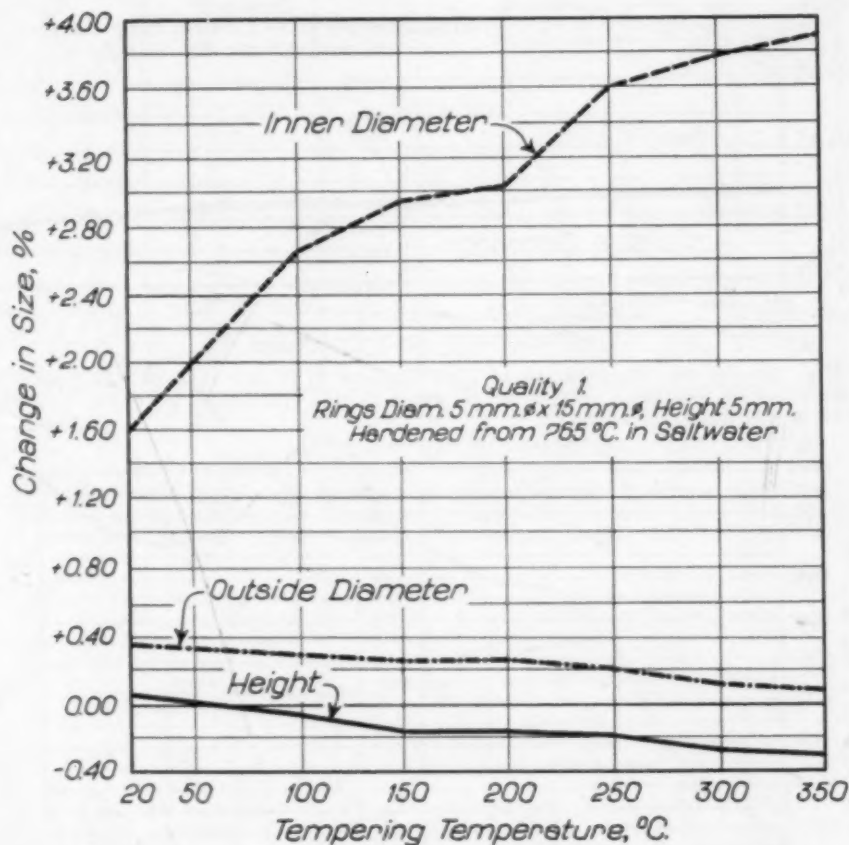


Fig. 29—Graphic Illustration of Size Change of Rings (Quality 1) 5 and 15 Millimeters in Diameter and 5 Millimeters Long, Hardened from 765 Degrees Cent. in Salt Water and Tempered.

such a deformation occur during the tempering that the sample has a tendency to approach the spherical shape, then the core would have been subjected to pressure and the surface to tension. If the deformation had occurred in the opposite direction, then the condition of strains would have been just the opposite. The above results have been studied under these assumptions, and the comparison has been

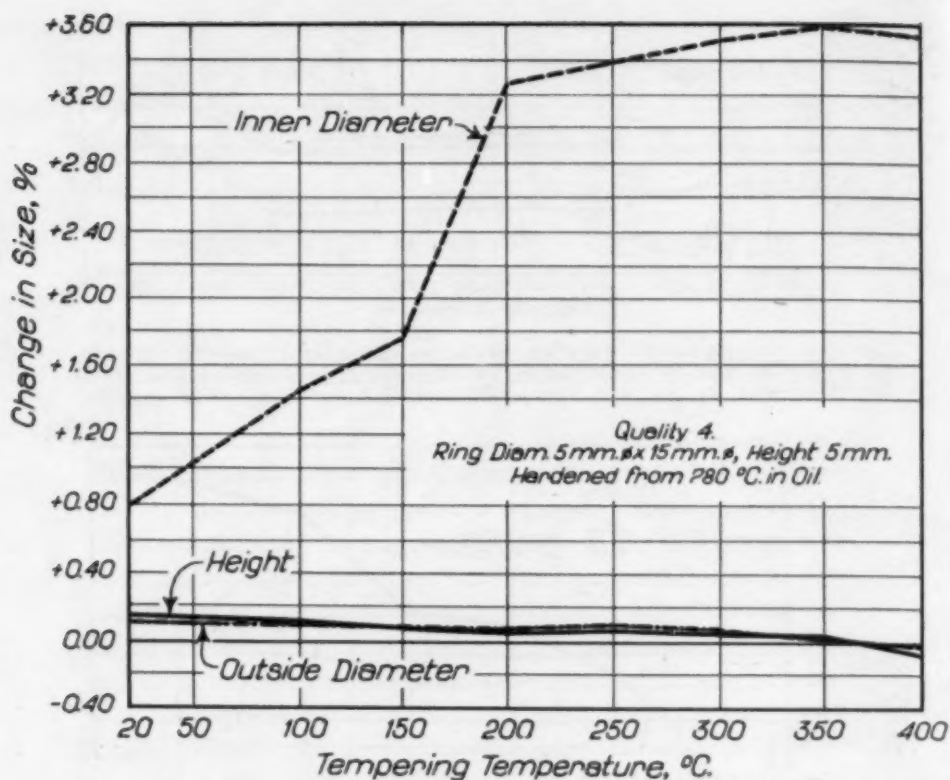


Fig. 30—Graphic Illustration of Size Change of Rings (Quality 4) 5 and 15 Millimeters in Diameter and 5 Millimeters Long, Hardened from 780 Degrees Cent. in Oil and Tempered.

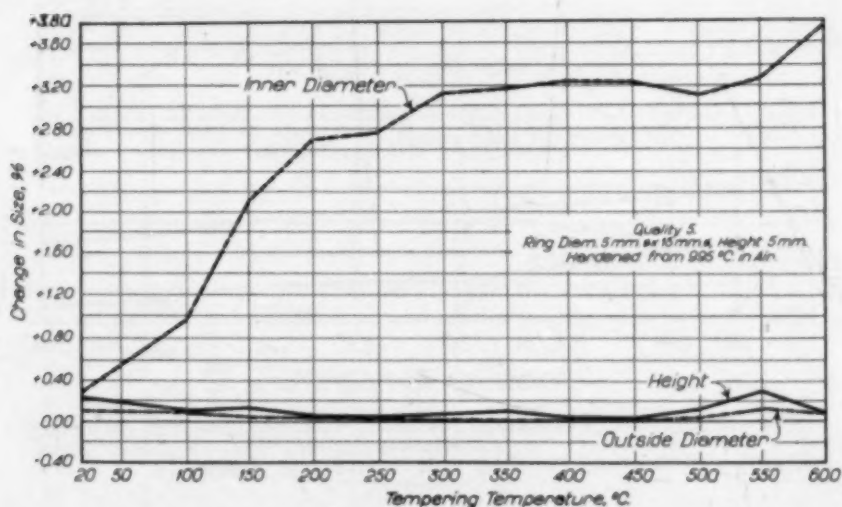


Fig. 31—Graphic Illustration of Size Change of Rings (Quality 5) 5 and 15 Millimeters in Diameter and 5 Millimeters Long, Hardened from 995 Degrees Cent. in Air and Tempered.

made for steels Nos. 1, 2 and 3 after tempering at 350 degrees Cent. (660 degrees Fahr.) and for steels Nos. 4 and 5 after tempering at 400 degrees Cent. (750 degrees Fahr.).

In Figs. 14 to 19 the diameter has decreased proportionately

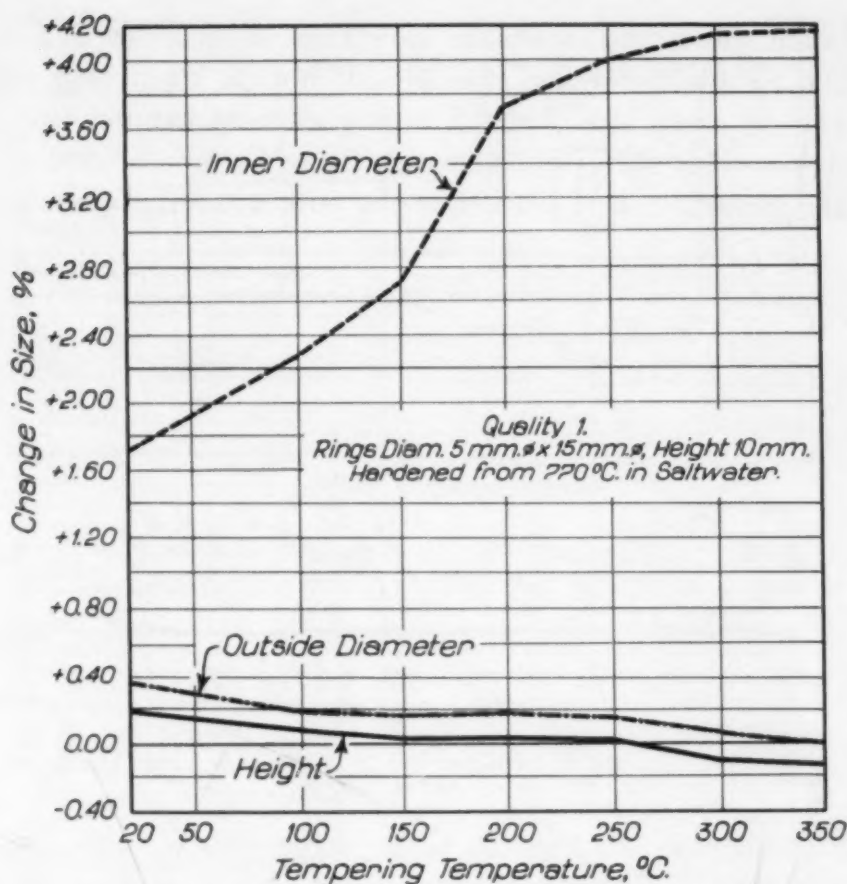


Fig. 32—Graphic Illustration of Size Change of Rings (Quality 1) 5 and 15 Millimeters in Diameter and 10 Millimeters Long, Hardened from 770 Degrees Cent. in Salt Water and Tempered.

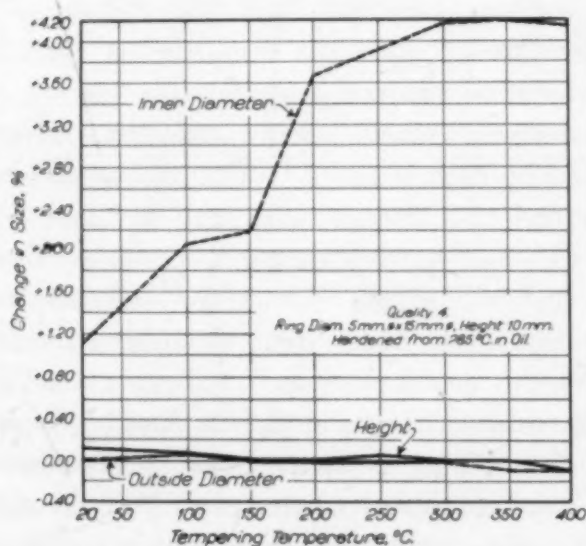


Fig. 33—Graphic Illustration of Size Change of Rings (Quality 4) 5 and 15 Millimeters in Diameter and 10 Millimeters Long, Hardened from 785 Degrees Cent. in Oil and Tempered.

more than the length, with the exception of cylinder 5 millimeter round x 15 millimeter for steel No. 1, normal quenching temperature, and for steels Nos. 3 and 5, high quenching temperature. These cylinders would, therefore, generally have been subjected to pressure at the surface and tension in the core after the quench.

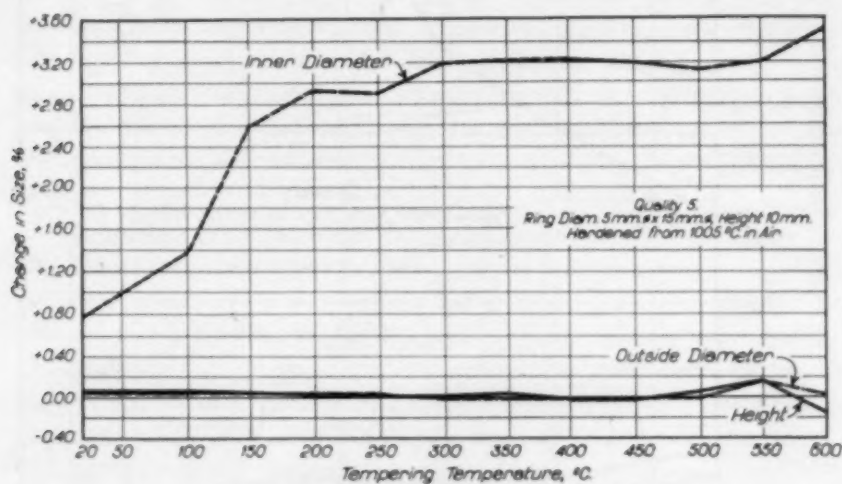


Fig. 34—Graphic Illustration of Size Change of Rings (Quality 5) 5 and 15 Millimeters in Diameter and 10 Millimeters Long, Hardened from 1005 Degrees Cent. in Air and Tempered.

With the samples, Figs. 20 to 22, which do not harden through, i.e., steels Nos. 1, 2 and 3, the length has decreased proportionately more than the diameter. This does not necessarily mean that the cylinders after quench have been under pressure in the core, but might also be explained by the fact that the outer hardened case shrinks with tempering, while the unhardened core does not. The core will, therefore, be subjected to pressure during tempering, and the surface to tension, which should create a deformation similar to the one caused by heat stresses from quenching. From the above it is also concluded that the stresses may increase when tempering a steel which does not harden through, provided that the drawing temperature is sufficiently low so that the plasticity of the material has not been increased to any extent in order to allow equalization of stresses. As the decomposition of the martensite accompanied by a decrease in volume occurs at a very low temperature, provided the time is sufficiently long, it seems possible that the stresses might increase on this account even when keeping at room temperature a quenched object, and it may also be that this condition is responsible at times for the cracking of quenched pieces after the quenching operation.

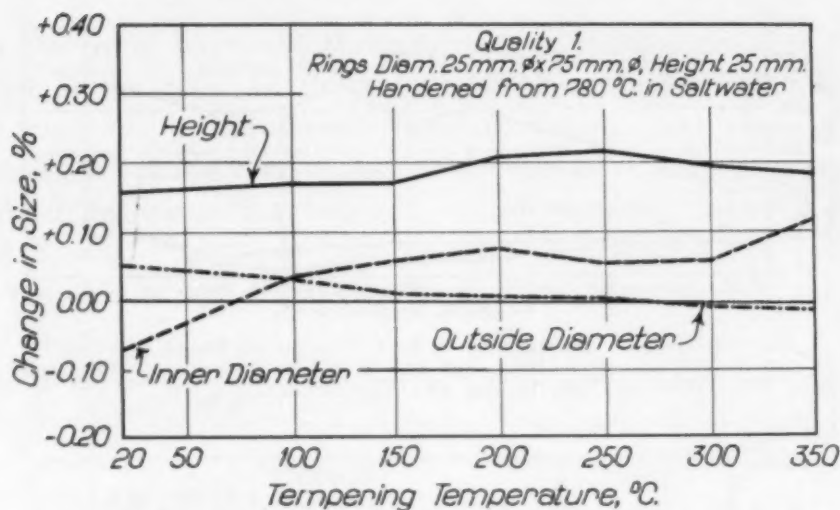


Fig. 35—Graphic Illustration of Size Change of Rings (Quality 1) 25 and 75 Millimeters in Diameter and 25 Millimeters Long, Hardened from 780 Degrees Cent. in Salt Water and Tempered.

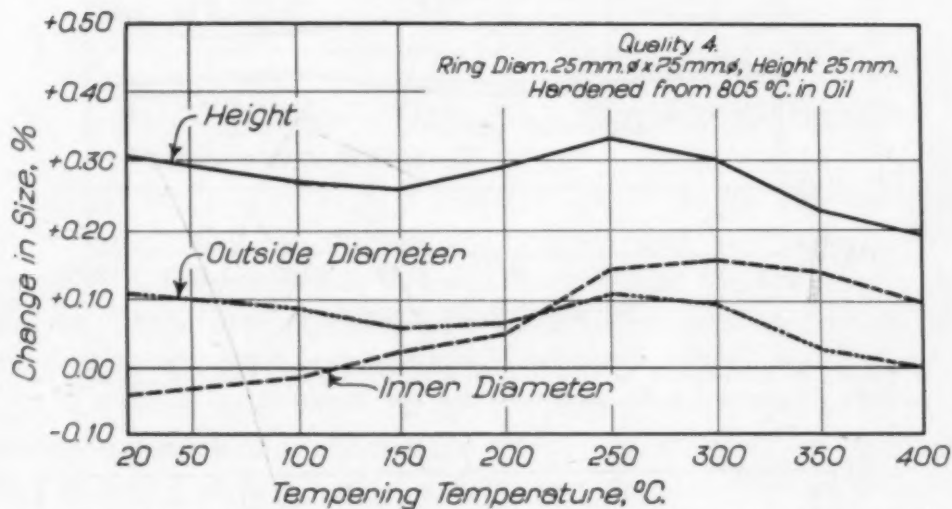


Fig. 36—Graphic Illustration of Size Change of Rings (Quality 4) 25 and 75 Millimeters in Diameter and 25 Millimeters Long, Hardened from 805 Degrees Cent. in Oil and Tempered.

With steels Nos. 4 and 5, with the exception of normal quenching temperature for steel No. 4, the length has decreased somewhat more than the diameter during the tempering. The surface would, therefore, have been under tension and the core under pressure.

With steels Nos. 1, 2 and 3, Figs. 23 to 25, the diameter has shrunk proportionately more than the height which in some cases even has increased. The disks must, therefore, have been subjected to great pressure at the core after quenching, which seems probable for the reason that the disks after the quenching have been deformed in much the same direction. The reason might also be that the disks

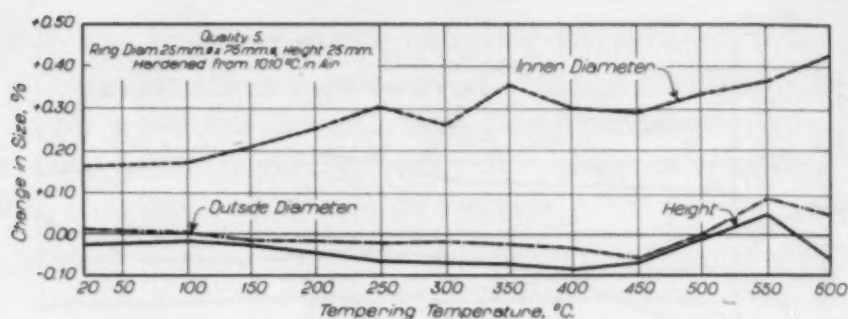


Fig. 37—Graphic Illustration of Size Change of Rings (Quality 5) 25 and 75 Millimeters in Diameter and 25 Millimeters Long, Hardened from 1010 Degrees Cent. in Air and Tempered.

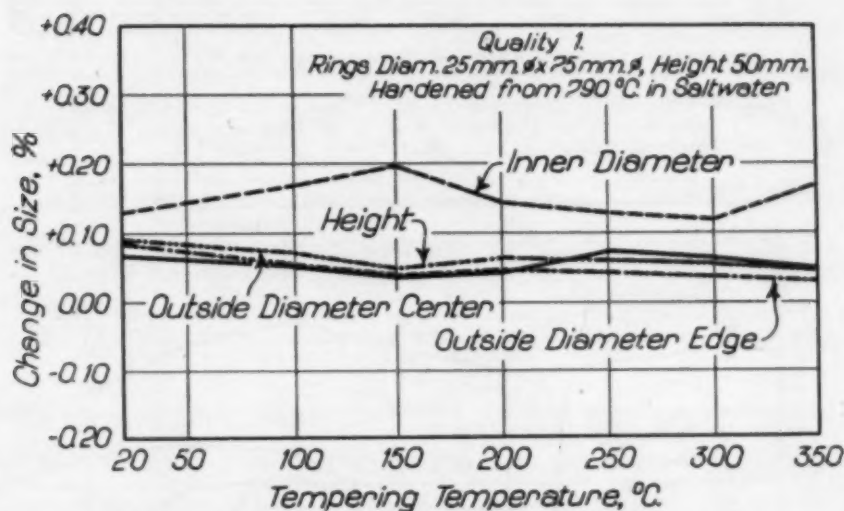


Fig. 38—Graphic Illustration of Size Change of Rings (Quality 1) 25 and 75 Millimeters in Diameter and 50 Millimeters Long, Hardened from 790 Degrees Cent. in Salt Water and Tempered.

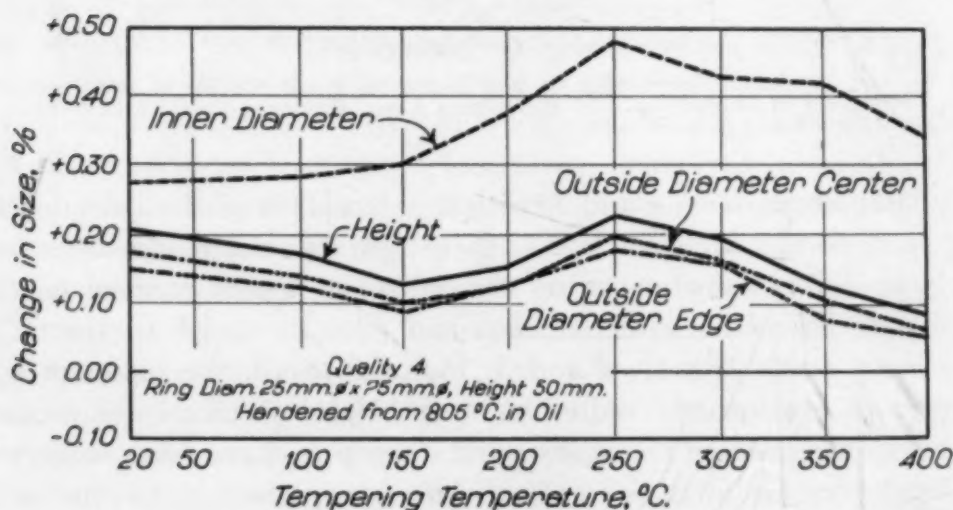


Fig. 39—Graphic Illustration of Size Change of Rings (Quality 4) 25 and 75 Millimeters in Diameter and 50 Millimeters Long, Hardened from 805 Degrees Cent. in Oil and Tempered.

are not entirely hardened through and, therefore, pressure has developed in the core, as the surface shrinks during the tempering. With steel No. 4, the diameter and the length have changed about the same, indicating that no larger stresses have been present. Steel No. 5 during tempering behaved entirely opposite to steels Nos. 1, 2 and 3, which conforms with the fact that, during the quenching, steel No. 5 changed differently from these other steels.

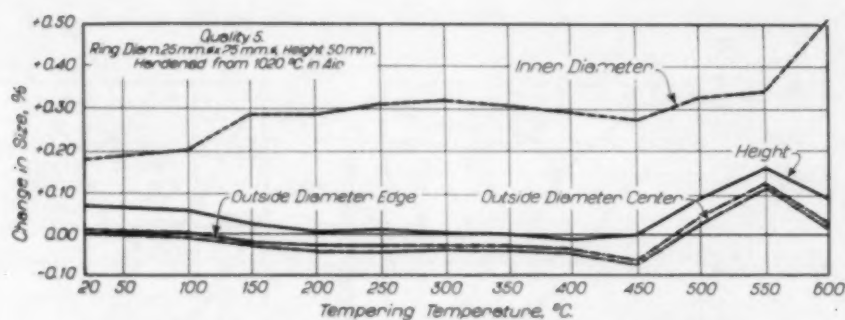


Fig. 40—Graphic Illustration of Size Change of Rings (Quality 5) 25 and 75 Millimeters in Diameter and 50 Millimeters Long, Hardened from 1020 Degrees Cent. in Air and Tempered.

With steels Nos. 1, 2, 3 and 4, Figs. 26 to 28, the differences are so small that no definite conclusions can be reached. With steel No. 5, the length has decreased more than the diameter. These samples, therefore, probably had been under pressure at the core and tension at the surface after quench.

With the small rings (Figs. 29 to 34) the inner diameter has increased considerably during the tempering, irrespective of the fact that it really should have decreased on account of the decreasing volume after the tempering. This proves that large stresses are present in the rings after the quench.

It is remarkable that the inner diameter is increased very rapidly with tempering up to 200 degrees Cent. (390 degrees Fahr.), which proves that a considerable equalization of stresses can occur at this low temperature. As a release of tension requires that the elastic limit of the material is lowered, this would evidently indicate that the elastic limit is reduced considerably when tempering quenched steels at a temperature of 200 degrees Cent. (390 degrees Fahr.).

The inner diameter of the rings of Figs. 35 to 40 has in all cases been increased. The condition is evidently identical with the small rings. The increase is not so great, however, as with the small

rings, which would indicate that the strains have not been so severe in the larger rings.

HEAT STRESSES DURING TEMPERING

When heating to and cooling from the tempering temperature new heat stresses may arise. When tempering tool steel it is, however, not necessary to consider that these stresses will create any per-

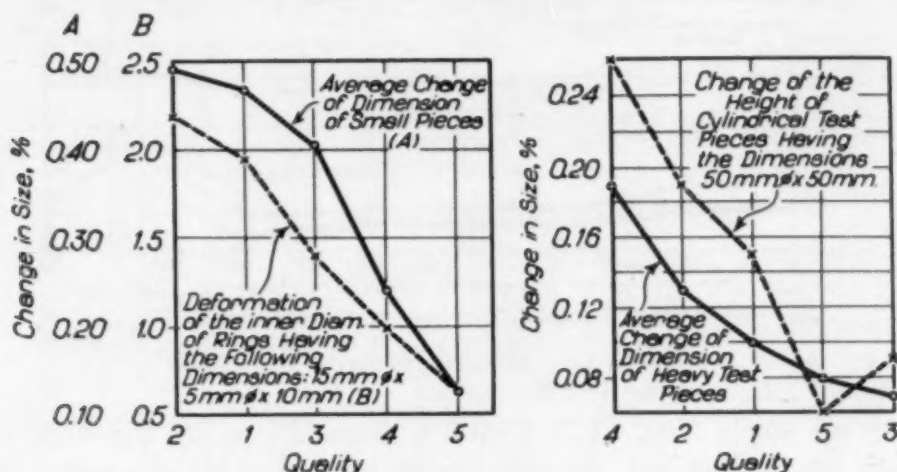


Fig. 41—Showing Agreement Between Dimension Change and Average Deformation of the Selected Test Pieces.

manent deformation, as the heating and cooling usually occur slowly, and the tempering temperatures generally are relatively low and, consequently, the plasticity of the material is small.

When tempering structural steels, however, the conditions are entirely different. Structural steels are often cooled rapidly after tempering for the purpose of avoiding blue brittleness. Large stresses might, therefore, be created which also has been proved by H. Bühler and Buchholtz (4). These steels generally are tempered at a relatively high temperature so that the elastic limit is lowered. Therefore, permanent deformation can be expected.

DECREASE OF DEFORMATION THROUGH TEMPERING

From a practical viewpoint, it is important to be able to equalize the dimensional change created during quenching by tempering. It is, however, an exception to accomplish a complete equalization, as it is very rare that all the dimensions at the same time would re-

gain their original lengths when tempered at a certain temperature; neither is it possible to deliberately select the tempering temperature on account of other important properties as would be desirable if only the dimension change was considered. It is possible to arrive at the best tempering temperature for various samples from Figs. 14 to 40, which show the influence of the tempering upon the dimension change. It should be emphasized, however, that when the tempering temperature is selected, it is necessary to decide which dimension is desired to show the smallest change, and set the temperature accordingly, for instance, with a cylinder 5 millimeter round x 15 millimeter, steel No. 1, Fig. 14, the diameter regains its original length at about 120 degrees Cent. (250 degrees Fahr.) while the length reaches the zero line at about 280 degrees Cent. (535 degrees Fahr.).

With steel No. 5, the unusual thing happens in that at three different tempering temperatures certain dimensions are able to regain the original length. This depends upon the expansion at the austenite decomposition and the contraction thereafter.

PRACTICAL TESTING OF THE DIMENSION CHANGE

From a practical viewpoint, it would be helpful to be able in some simple way to judge the tendency of a steel for dimension change during quenching. Such an estimation is not obtainable, as has been emphasized previously, by determining the change of specific gravity during quenching. In order to obtain an entirely correct idea of the tendency of a steel to change during hardening, the only way would be to quench numerous samples and measure the dimension changes in various directions. Such a procedure, however, is not suitable for practical work, as it is tedious, expensive and requires a great deal of material for samples. It has, therefore, been very common that the tests have been limited to one certain type of sample, for example, a cylinder, a disk, or some special shape, the dimensional changes of which have been measured after hardening. The sample piece has usually been selected at random, and the results obtained could hardly be considered highly decisive. It has, therefore, been considered important to determine which sample has in the present investigation given a deformation closest resembling the average deformation of the steel and which, there-

fore, would be most suitable as a standard sample for practical investigation of dimensions.

The requirement of a suitable test piece is that the dimensional change should be as large as possible in order to increase the accuracy of the measuring. It has not been considered quite so important, however, that the same absolute measurement is obtained as with the average dimension change if only the relative comparison between the steels remains the same. For comparison, differentiation has been made between smaller and heavier dimensions in order to take into consideration the tendency of the steels to harden through.

From this comparison it has been found that the inner diameter of the small rings, dimension 15 millimeter round x 5 millimeter round x 10 millimeter, gives the best indication of the dimension change of the smaller samples, and that the height of the cylinders, dimension 15 millimeter round x 50 millimeter, gives the best indication of the dimension change of the larger test pieces. The agreement between the dimension change and the average deformation of the selected test pieces is shown in Fig. 41.

CONCLUSION

The dimension changes resulting from quenching and tempering have been studied on a number of samples of five different tool steels. For comparison, samples of soft iron have also been included. The principal results of the present investigation can be summarized as follows—

The deformation which occurs during the heat treatment of steel is caused by heat stresses and also by stresses which occur in connection with the structure transformation.

The heat stresses seek to deform the hardened object so as to give it a shape approaching the sphere. The transformation stresses tend to deform the sample in the opposite direction. Which of the stresses that are to determine the character of the deformation depends upon the composition of the steel, shape and size of sample, quenching temperature, etc. It is, therefore, difficult to determine beforehand how an object is going to be deformed during heat treating.

Factors which increase the temperature variations during heat treatment, such as rapid velocity of cooling, decreased heat conduc-

tivity, increase the deformation on account of heat stresses. Factors which increase the elastic limit of the material at elevated temperature reduce the deformation on account of heat stresses. Transformation stresses decrease, in case part of the structure is a constituent which does not participate in the transformation; for example, retained austenite or excess carbide.

With samples that harden all the way through, steel No. 5 has given approximately four times less deformation, and steel No. 4 about two times less deformation than steels Nos. 1, 2 and 3, which latter are about equal. With larger samples, where the steels Nos. 1, 2 and 3 do not harden through, the condition is different. Under this condition, steel No. 3 has given as good a result as steel No. 5, and steels Nos. 1 and 2 are only slightly inferior. Steel No. 4 has, however, given considerably larger deformation than the other steels. In the steels Nos. 1, 2 and 3, which have not hardened through, the deformation is decreased with reduced hardness penetration.

The steels considered as nondeforming of the type of steel No. 4 are usually, from the point of view of deformation, only to be preferred to carbon steel and steel of the type of steel No. 3 where the dimensions are such that the latter steels harden all the way through. Steel No. 5 is a steel, however, with especially small deformation in both small and large dimensions and may, therefore, be considered as the most all around steel, as far as permanent shape is concerned.

In order to obtain the least deformation of a tool steel, it should have the following qualities:

1. It should allow the slowest possible cooling velocity during quenching.
2. A relatively large amount of austenite should be retained without impairing the hardness.
3. The amount of carbide not participating in the transformation during quenching should be large.

An increase of the quenching temperature does not necessarily increase the deformation because the increased heat stresses might be compensated by the transformation stresses. When tempering hardened objects, the dimensions are changed on account of structure transformations and also on account of release of strains. When tempering at high temperature with rapid cooling, deformation might also develop on account of heat stresses.

Strain release seems to occur rapidly up to about 200 degrees Cent. (390 degrees Fahr.) and thereafter slower. By studying the dimension change during tempering, it is possible to obtain some idea concerning the stresses after hardening.

Only in exceptional cases can we count upon complete restoration of dimensions by tempering. From the experiments just described, it is possible to determine which drawing temperature or temperatures give the best results with the test pieces and steels we have selected.

In order to find some simple way to determine the tendency of a steel to change dimension during hardening, it would seem best for the small size samples to measure the dimension change of the inner diameter on a ring about 15 millimeter round x 5 millimeter round x 10 millimeter, and for the large size samples to measure the dimension change of the height of a cylinder about 50 millimeter round x 50 millimeter.

The author wishes to express his appreciation to Uddeholms Aktiebolag for its co-operation and for the permission to have the present paper published; and also to Dr. Ragnar Arpi, Chief Metallurgist of Uddeholms A/B, upon whose initiative this investigation was started, and who has generously given his assistance in its preparation.

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ERRATUM

On page 452 of the June 1939 issue of *TRANSACTIONS*, Volume 27, there appears an editorial error in F. B. Foley's discussion of Cramer and Bast's paper entitled "The Production of Flakes by Treating Molten Steel with Hydrogen and the Time of Cooling Necessary to Prevent their Formation."

The third paragraph reads:

"Alloy steels, when made in a basic furnace, are much less likely to develop thermal cracks than a similar steel made by the acid process."

This should be corrected to read:

"Alloy steels, when made in an acid furnace, are much less likely to develop thermal cracks than a similar steel made by the basic process."

In order that members may have the correct version of this discussion, the following two pages may be substituted for those already printed in the June 1939 issue of *TRANSACTIONS*, Volume 27.

electrodes. Apparently the hydrogen in solution would be unusually high in these heats and would keep coming out of solution while the steel was still in the mushy stage so that the expansion of the trapped hydrogen would be sufficient to cause the wildness in the mold. Similar difficulties can be traced to wet limestone, wet hot tops and wet stool brick. Some furnaces which are fired with steam atomization of the fuel occasionally have the same difficulty.

About 20 years ago a paper was given by Styri, in which he connected the cause of wildness, etc., in ingots to atmospheric conditions, and I have noticed in steel castings that there is an increase in the amount of porosity during warm humid weather. Whether hydrogen caused this I do not know, but I do believe while it is a contributing factor in the formation of flakes it is hard to prove that it is the primary cause. I do not know whether the observation that moisture is the cause of porosity in castings is universal in its application, but it has been made often enough to indicate that hydrogen is an important factor in the soundness of the castings, whether it be a commercial casting or an ingot.

JOHN CHIPMAN:³ I would like to remark that this is a very beautiful piece of work and quite different from the older work that was reported on gases in metals some 10 or 20 years ago, in which almost everything that could go wrong with the metal was blamed on gas. Here we have some very definite, tangible evidence of a perfectly well recognized effect of hydrogen gas. This is finally something upon which one can place his finger and say here is an effect of the hydrogen.

F. B. FOLEY:⁴ The name "flake" as applied to the type of defect discussed in this paper by Cramer and Bast is a contraction of the word "snow-flake" used during the war period of 1917-18 to describe the appearance of a defect found in tensile test specimens and ordinary fractures from certain nickel steel gun forgings. Such fractures contained bright areas such as might be produced by pressing a small piece of thin silver foil onto the fracture. Tensile test bars found to contain these defects developed full tensile strength and elastic limit, but were characterized by low elongation and reduction of area. Another type of defect, which undoubtedly was a thermal crack and produced by too rapid cooling immediately following the forging, had a different appearance, being very coarse where the crack existed, and gave results in tensile tests which were low in every respect. It is this latter type of defect with which I believe the authors are at present concerned and which are aggravated by the presence of hydrogen and readily developed by too rapid cooling following hot working.

It is characteristic of cooling cracks that they are confined to the deep-seated layers of metal and do not occur near the surface nor close to the ends of bars or forgings, yet it is well known that, in spite of the fact that they are rarely if ever found in the portions of the metal where cooling is most rapid, rapid cooling develops them and slow cooling after forging or rolling prevents their occurrence. It was my observation of thermal cracks which I examined during the war period that they were generally associated with a coarse austenitic grain size in the forging.

³Department of metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

⁴Superintendent of research, The Midvale Company, Nicetown, Philadelphia.

Another fact which seems to run counter to the general idea that it is wholly the cooling rate which causes cooling cracks is that they rarely if ever occur in small sections, under say $2\frac{1}{2}$ inches square, which cool more rapidly than the larger sizes. Of course it can be said that the requisite stresses are not present in the small sections.

It is my experience that once a steel, subject to thermal cracking during cooling, has been cooled from the forge or mill in such a manner as to avoid their presence it may then be reheated to a normalizing temperature and cooled in air, a rate which if applied to the piece as it comes off the mill or hammer would cause cracks to develop, without developing any cracks. Hultgren found that chromium-bearing steel finished at a low temperature in hot working was generally free of cooling cracks, although the same steel finished hot readily developed them.

Alloy steels, when made in an acid furnace, are much less likely to develop thermal cracks than a similar steel made by the basic process.

C. E. SIMS:⁵ After the authors' presentation there can be very little doubt as to the cause of this trouble, but there does seem to be some question as yet as to just what the mechanism is. I would like to cite a few experiences which I have had in cast steel which has an analogous condition and which is most certainly due to hydrogen.

We find in normalized test bars, that is, normalized as 1-inch sections, after breaking in a tensile machine, certain little spots usually about $\frac{1}{16}$ of an inch to as big as $\frac{3}{16}$ of an inch in diameter. When the bar is broken the fracture will usually zigzag back and forth through those spots resulting in a very jagged fracture. At these peaks and valleys you will find an irregular spot usually fairly round and in the center of which there will be a very tiny cavity. This spot will have a very fine-grained fracture like a hard, brittle steel, very much like a high carbon steel, whereas the rest of the area will have a fibrous appearance. This brittle area is always concentric with the small cavity.

Now, if we should age that bar at some low temperature, even at room temperature if it is aged long enough, and then break it, the spots will not appear. This is evidence that there are no fissures or cracks prior to breaking the bar and this condition is due to the manner in which the steel fractures. At room temperature it may require two or three weeks but if we age it at 200 or 300 degrees for about 15 hours the spots will disappear. At temperatures of 500 to 700 degrees the condition is cured in a very few hours. A peculiar thing about it is that, if the bar is renormalized after aging to remove the spots and then pulled without reaging, the spots would reappear. This condition in cast steel has been known and worried about for a good many years. Although hydrogen had been suspected, the first definite indication was obtained when we broke some welded bars that had been welded with atomic hydrogen. Of course, the weld metal all solidified in an atmosphere of practically pure hydrogen and we found the same condition but in a greatly exaggerated state. Instead of spots $\frac{1}{16}$ inch in diameter there were spots $\frac{1}{8}$ inch

⁵Research metallurgist, Battelle Memorial Institute, Columbus, O.

